Effect of Thermomechanical Treatment of Medium Carbon 49MnVS3 Microalloyed Steel on Mechanical Properties

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February, 2001

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A Thesis Submitted
in Partial Fulfillment of the Requirements
for the Degree of

Master of Technology

by

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to the

Department of Meterials and Metallurgical Engineering
Indian Institute Of Technology
Kanpur

February, 2001

Dedicated

To

My Parents

Certificate

This is to certify that the thesis entitled, **'Effect of Thermomechanical Treatment of Medium Carbon 49MnVS3 Microalloyed Steel on Mechanical Properties'**, is the record of the work carried out by **Guru Prasad Dinda** under my supervision and has not been submitted else where for the award of the degree

(Prof. S. Bhargava.)

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Acknowledgements

I would like to express my deep sense of gratitude and indebtedness to **Prof. S. Bhargava** for his valuable guidance, encouragement and inspirational discussion throughout the course of this investigation It has truly been a learning experience to associate with him and work towards the completion of this work

I am very highly indebted to **Dr. M.N.Mungoli** for his invaluable help. I express my sincere thanks to **Mr. B.K. Jain**, lab in-charge Materials Testing lab, for his invaluable help in carrying out tensile tests and micro hardness measurements

I am grateful to **Kumar ji, Malhotra ji Mr.B.Sharma, Mr. Agnihotri** and all staffs of MME dept. for their assistance in the experimental work.

Last but not least, I would like to acknowledge the affection, respect and support received from my friends **Shantanu, Dulal, Projit, Mahato, Sanjay, Jain, Arrora, Srinivas and Binod.** Their company has made my stay in IITK a memorable one.

Guru Prasad Dinda

Abstract

In drop forging of parts for the transport industry the classical quench and tempering of alloyed steels is nowadays substituted by direct continuous cooling of microalloyed steels with elimination of quench cracking and expensive strengthening and stress relieving cycle. Nevertheless, there are some limitation on strength and toughness by this technique. The aim of the present study is to suggest general guideline for the design and deformation patterns and cooling conditions of microalloyed steel (49MnVS3), to be used in automobile industry, which has properties comparable to quenched and tempered variety.

In the present investigation the steel samples were rolled at two-phase field i e at 760°C and single–phase field i e at 800°C, 900°C, 1000°C, 1100°C 1200°C respectively followed by either quenched in water or cooled through two stage cooling (TSC). In addition, a set of samples was subjected to further tempering after water quenching/TSC By using a two-step cooling formation of a ductile ferrite is promoted in the first (slow) cooling step (on quite air), followed by water quenching in a second step generating hard martensitc-bainitic secondary phase components.

The treatment involving hot rolling at 760°C followed by quenching or TSC provides a dual-phase structure comprising of polygonal ferrite and martensite. After tempering the structure comprises of ferrite and tempered martensite. However, under these conditions the steel shows a lower yield strength (~550 MPa) than the highest achievable value (~1030 MPa) in the same steel when rolled at single-phase field and cooled through TSC. However, the ductility of the steel under this state is considerably superior than the steel rolled at single-phase field.

The route involving hot rolling of the steel in the single phase austenite field in the temperature range from 800°C – 1200°C followed by the quenching and tempering does not provide much variation in microstructures and hence the consequent properties. The route involving hot rolling of the steel in the single-phase austenite field in the temperature range from 800°C – 1200°C followed by two stage cooling and annealing provides considerable variation in its microstructure and mechanical properties of the steel

The present investigation shows that the steel, hot rolled at 800°C – cooled through TSC – annealed at 420°C, consists of a small volume fraction of polygonal ferrite and bainite as the matrix. In contrast, steel hot rolled at 1200°C – cooled through TSC – annealed at 420°C consists of a somewhat smaller volume fraction of polygonal ferrite and tempered martensite as the matrix The steel rolled in between these two temperatures, i.e. at 900°C, 1000°C and 1100°C, shows polygonal as well as elongated ferrite and a mixture of bainite and tempered martensite as the matrix. The volume fraction of the bainite in the matrix, however, decreases as the rolling temperature of the steel increases. Among the thermomechanical routes investigated in the present study, it has been found that the treatment involving a combination of hot rolling followed by two stage cooling and annealing is the best thermomechanical route for the processing of 49MnVS3 micro-alloyed steel from the point of view of optimizing its mechanical properties It was also observed the optimum mechanical properties are obtained of the TSC + annealed material when rolled at temperature between 900°C to 1000°C.

CONTENTS

CHAPTER 1

INTRODUCTION

The development of high strength materials, coupled with good formability, has always been the aim of material scientists Alloying, mechanical working, heat treatment, grain size control and nuclear irradiations are some of the techniques, which may be taken recourse to individually or in combination for this purpose Individually, each one of them has a limited strengthening effect In order to attain enhanced properties, various combinations of these unit operations may be adopted advantageously. Thermomechanical treatment is one such combination As the name suggests, it is a combination of heat treatment and mechanical working. Any combination of heat treatment and mechanical working cannot be termed a thermomechanical treatment. The term thermomechanical treatment refers to that treatment in which plastic deformation is carried out in such a way that phase transformation is affected by it.

At the beginning of the 70's, the introduction of the vanadium microalloyed medium carbon steels designed for direct cooling after forging was an excellent example of invention in steel industry. A review of the further development of such new steel is given in Fig. 1.1[Kaspetr et al., 1997]. The carbon content has been progressively reduced (from 0 49% in 1971 to 0.27% nowadays) in order to improve the toughness through a large ferrite fraction. Microalloyed steel has proven to be a very important material as it can provide a higher strength with smaller alloying additions and less heat treatment expenses. Quenching followed by tempering (QT) has been the conventional route but it requires expensive strengthening and stress

relieving cycles and also has product rejection. The attempt to overcome these difficulties led to the introduction of continuous cooling (CC) of medium carbon microalloyed steel from the working temperature producing a ferrite-pearlite microstructure in the products But there is hardly any cooling rate using CC by which mechanical properties could be achieved close to those after QT of the same steels

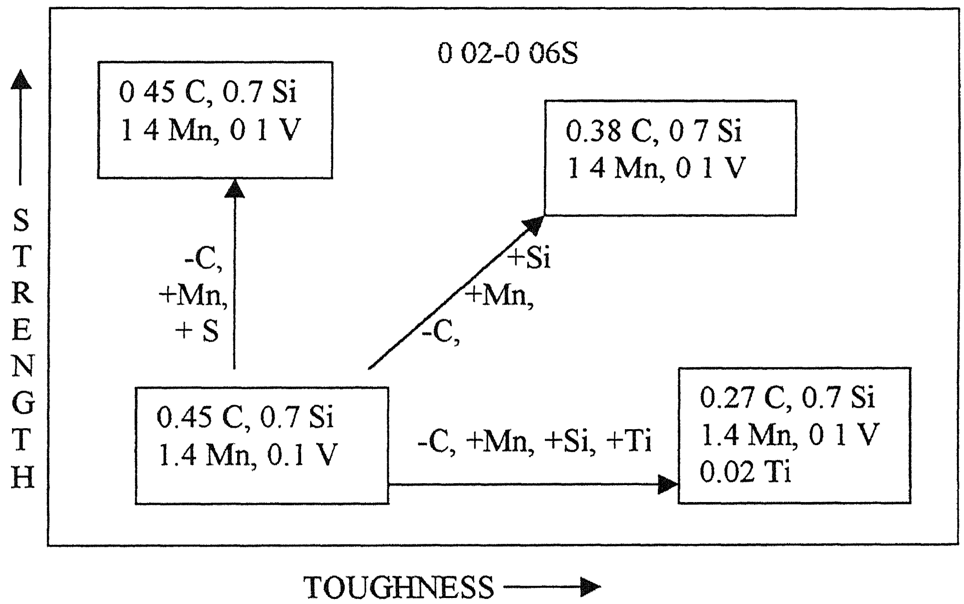


Figure 1.1 Trends in the development of microalloyed ferritic-pearlitic steels designed for direct cooling to improve final mechanical properties.

1.1 Two-Step Cooling (TSC): An Alternative to QT-Route

The aim of the new approach is to produce a ferrite-bainite/martensite microstructure after direct cooling from the rolling temperature, consisting of homogeneously distributed ferrite fraction combined with a hard second phase. For this purpose rolling temperature must be lowered to enhance ferrite nucleation due to a large specific area of austenite grain structure. This leads to a higher and more homogeneously dispersed ferrite fraction. The principle of the so-called two-step cooling (TSC) after rolling, compare to CC, is given in figure 1.2 [Kothe et al., 1997].

This new cooling design promotes the ferrite formation during the slow first –step cooling with a cooling rate T_1 (natural cooling on still air) It occurs at rather high temperatures with two consequence firstly, the ferrite is polygonal (good toughness) with a high carbon content (high strength), secondly, the carbon is able to diffuse from the supersaturated ferrite into the adjacent austenite, the latter is being stabilized promoting the formation of bainite or even martensite instead of pearlite. Accelerated cooling (water quenching) in the second step with a cooling rate T_2 leads to a bainite/martensite transformation of the remaining austenite providing a hard second phase.

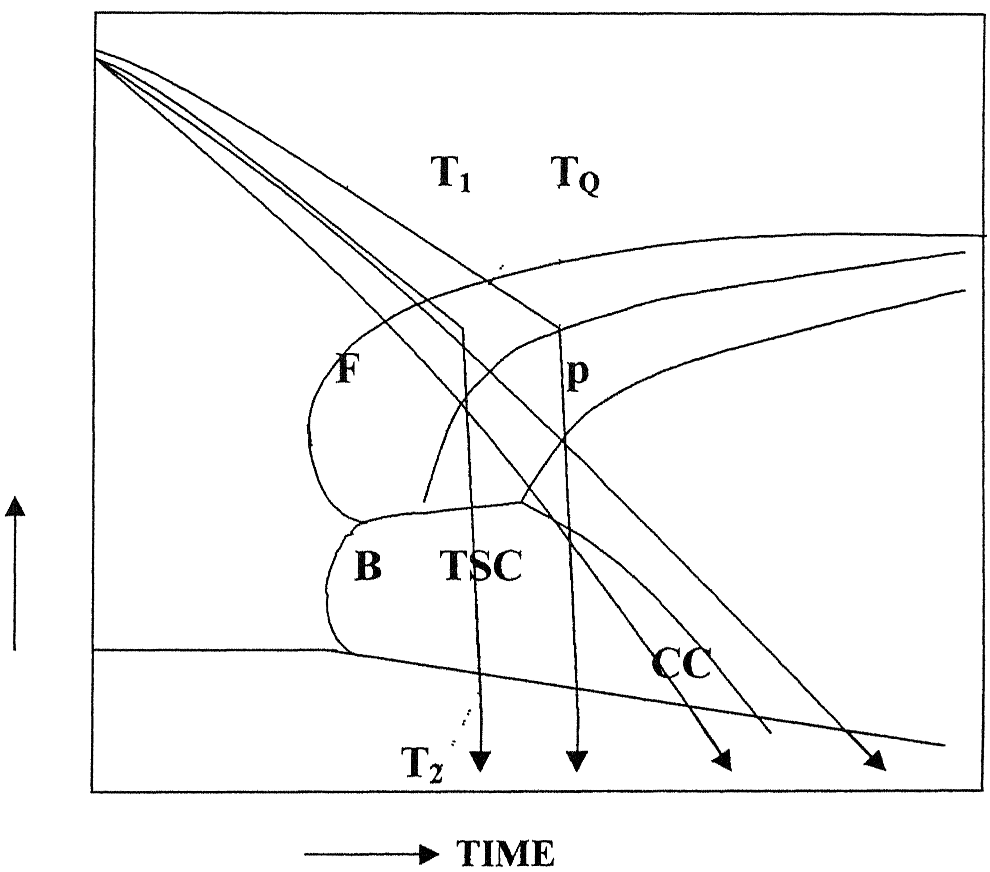


Figure 1.2 Processing schedule for two-step cooling (TSC) after rolling showing the variation of quench temperature T_Q; comparison two this continuous cooling (CC) with varying cooling rate.

CHAPTER 2

LITERATURE REVIEW

INTRODUCTION

2 1 1 The rationale of Microalloyed High Strength Steels

The motivation for the use of high strength low alloy (HSLA) steels is cost reduction Higher strength steels can sustain service loads over thinner sections, thus providing weight savings and use of less steel Lower costs can also be realized if microalloyed steels can replace alloy steels containing significant amounts of expensive elements like Nickel, Chromium or Molybdenum However, the most significant cost reduction provided by microalloyed HSLA steels is the elimination of expensive heat treatment and their use in the mill processed condition, as-rolled, which provides considerable energy saving As-rolled microalloyed HSLA steels exhibit properties which are comparable in many ways to those of normalized or quenched and tempered products, yet avoid the cost of heat treatment, handling, energy decarburization, scale loss (oxidation) etc

In terms of mechanical properties, the heat-treated (quenched and tempered) low alloy steels offer the best combination of strength and toughness (Table 2 1) [Metals Handbook, 1990] However, these steels are available primarily as bar and plate products and only occasionally as sheet and structural shapes In particular, structural shapes (I-beams, channels, wide flanged beams, or special sections) can be difficult to produce in the quenched and tempered condition because shape warpage can occur during quenching Heat treating steel is also a more involved process than the production of as-rolled steels, which is one reason why as-rolled steels are an

Table 2 1 General comparison of mild steel with various high strength steels[Metals Handbook, 1990]

	Chemical Composition(%) [a]						
Steel	C (max)	Mn (max)	Si (min)	Other	Min Yield Strength(Mpa)	Min UTS (Mpa)	Min Ductility[b] (%)
Low-Carbon steel	0 29	0 60	0 15	(c)	170-250	310-445	23-30
As-hot rolled C-Mn steel	0 40	1 00	0 15		250-400	415-690	15-30
HSLA steel	0 08	1 30	0 15	0 02 Nb	275-450	415-550	18-24
Heat treated C steel Normalized[c]	0 36	0 90	0 15		200	415	24
Quenched and Tempered	0 20	1 50	0 15	5 ppm B	550-690	660-760	18
Q and T low-alloy steel	0 21	0 45	0 20	0 45 Mo, 50 ppm B	620-690	720-800	17-18

[a] Typical compositions include 0 04% P (max) and 0 05% S (max)
[b] Elongation in 50 mm gage, [c] If copper is specified, the minimum is 0 20%

attractive alternative The as-rolled HSLA steels are also commonly available in all the standard wrought forms (sheet, strip, bar, plate and structural shapes)

In general the rationale behind the development of these steels can be summarized as

- Filling the yield strength gap between the simple carbon and mild steels with σ_y~300 MPa and heat-treated low alloy steels with σ_y~850 MPa with a mill processed microalloyed steel not requiring a separate heat treatment operation
- Producing high yield strength so that a greater load bearing capacity is possible with thinner sections
- Decreased processing costs, and obtaining a higher yield strength material, by using mill-processed steel with consequent energy conservation
- A minimal use of expensive and scarce alloying addition, with the consequent conservation of strategic scarce materials
- High weldability
- A high resistance to brittle cleavage and low energy ductile fracture, and a low ductile-brittle transition temperature
- An ability to cold form to some extent, particularly by bending and good ductility and toughness through the thickness of rolled strips and plates
- Optimal property combinations per unit cost

These requirements have been met by three main classes of material microstructures

Steels comprising ferrite – pearlite microstructures, which will be the main subject of this discussion Acicular ferrite structures, which may be considered to be very low carbon variants of bainitic steels developed in the early 1950's These are less popular

than the ferrite-pearlite structures, largely due to increased cost Dual phases microstructures, which were developed specifically to increase the formability and aimed particularly for use in the manufacture of automobile components

2 1 2 Applications of HSLA Steels

HSLA steels were first used as structural shapes and plates in the early 1960's because of their ability to be welded with ease By the early 1970's, they were also used in pipelines at both elevated temperature and severe arctic conditions Later in the 1970's, concurrent with the energy crisis, another dominant application involved the use of HSLA steels to reduce the weight of parts and assemblies in trucks and automobiles In the 1980's, bars, forgings and castings have emerged as applications of particular interest Shapes such as elbows and fittings for pipelines are also being cast out of microalloyed steel HSLA steels are used in a wide variety of applications, and their properties can be tailored to specific applications by a suitable combination of composition and microstructures obtained by processing in a mill For example, low – carbon and closely controlled carbon equivalent values provide good toughness and weldability Good yield strength and fracture toughness results from a fine grain size

Oil and gas pipelines While tensile strength is a key requirement in pipelines, other properties are no less critical for the fabrication and operation of oil and gas pipelines They include weldability, fracture toughness and corrosion resistance, which are met by HSAL steels

Automotive applications Experience in the application of cold- and hot-rolled HSLA sheet in automotive applications indicates the importance of requirements with respect to stiffness, crash behavior, fatigue life, corrosion resistance, acoustic properties and of course, formability and weldability In addition to improvement in

mileage per unit fuel consumption through weight reduction, benefits may be found in increasing payloads without a change in fuel consumption Candidate applications here include trucks, rail cars, off-highway vehicles and ships Ship applications are however limited by buckling and stiffness considerations

Offshore applications The essential characteristics of steels for these applications include

- Yield strength in the regions of 350 to 415 MPa
- Good weldability
- High resistance to lamellar tearing
- Lean compositions to minimize preheat requirements
- High toughness in the weld heat-affected zone
- Good fracture toughness at the designated operating temperatures

Some of these goals have been realized through a reduction in impurities such as sulphur, nitrogen and phosphorus in the steelmaking process Controlled rolling and accelerated cooling of niobium steels have allowed a reduction in carbon contents, which is important to enhance weldability Modification of sulfide inclusions is done by additions of rare-earth elements or calcium to form spheroidal inclusions This approach usually results in both the elimination of lamellar tearing and an improvement in transverse impact properties Bars, forgings and castings have emerged as the latest area of opportunity for microalloyed HSLA steels, and there has been a rapid assimilation of technologies developed in other product areas In particular, the search is on for higher strengths, elimination or reduction of heat treatments, simplified cold finishing or machining operations, and improved toughness, machinability and weldability

2 2 Growth of Microalloyed Steels

Starting with plain carbon steels containing 0 3 wt % C, little importance was attached to yield strength, toughness or weldability of HSLA steels, design being based on tensile strength This was standard steel in the late 1930's

2 2 1 Grain Refined Steels

By the early 1950's the beneficial effect of ferrite grain refinement on yield strength and the ductile-brittle transition temperature had been demonstrated [Petch, 1953, Cracknell and Petch, 1955] Initially, grain refining used Al-N additions [Weister and Ulmer, 1959] Later other grain refining additives like niobium, titanium or vanadium also were found to contribute to precipitation strengthening and higher yield strengths [Thomson and Krauss, 1989, Irvine et al 1970]

2 2 2 Precipitation Strengthening

It was found that grain refinement could take place in niobium steels, but no precipitation strengthening was present if the normalizing or austenizing temperature was a conventional $Ac_3 + 30\ ^0C$ With higher normalizing temperatures, precipitation strengthening became progressively greater but only at the expense of a pronounced loss in toughness On the other hand, vanadium steels showed precipitation strengthening when normalized from conventional temperatures and the impact toughness was greater than in the niobium steels [Honeycombe, 1986]

2.2.3 Development of Controlled Rolling

The use of controlled rolling at a low finish rolling temperature is to produce fine recrystallized or elongated unrecrystallized austenite grain morphologies, which

then transforms to fine polygonal ferrite on cooling to give excellent yield strength and associated toughness [Irvine et al , 1970, Phillips and Chapman, 1966, Duckworth, 1965] Much detailed research to understand the mechanisms present during controlled rolling has been carried out [Tanaka, 1981, Tamura et al , 1988] Conventional controlled rolling was able to increase the yield strength to 450-525 MPa with the impact transition temperature (ITT) as low as –80 ^{0}C

2 2 4 Development of Controlled Cooling

The development of controlled cooling, i e , accelerated cooling of either a plate, or more particularly of hot rolled strips, followed quickly the development of controlled rolling The reason for this was that the ferrite grain size decreases with decreasing austenite to ferrite transformation temperature with a consequent improvement in both yield strength and toughness (lower impact transition temperature, ITT) Accelerated cooling of plate materials can be carried out by air circulation, water sprays or mist cooling, but care is necessary to prevent excessive surface cooling and the formation of matensite/bainite structure Lamellar flow cooling is widely used for accelerated cooling of hot strip mill products [Sigalla, 1957]

2 2 5 Low carbon ferrite-pearlite Steels

In the 1960's the use of exceptionally low carbon pearlite free steels was considered The economic and production difficulties with these steels led to somewhat higher carbon contents of 0 03 to 0 08 wt %, the pearlite reduced steels These steels produced yield strengths of 550 MPa and an ITT of – 70 $_{0}$C [Phillips and Chapman, 1966]

2 2 6 Acicular Ferrite Steels

In the 1950's, low carbon (0 12/0 15 wt%) bainitic steels were developed on a commercial scale The proof stress values of 450-900 MPa, depending on the carbon content and transformation temperature, had an advantage over the ferrite-pearlite HSLA steels, but the toughness was considerably inferior so that interest in bainitic steels waned a ferrite-pearlite steels of lower alloy content and good strength – toughness combinations were developed The work of Coldren et al (1978) led to commercially attractive alloys based on 0 03 wt% C, manganese-molybdenum-niobium that were produced as hot rolled strips for line pipes This steels was produced with fully acicular ferrite bainite structures, with a modest yield strength of about 500 MPa and a very low ITT

2 2 7 Dual phase steel

HSLA steels showed low stretch formability The problem was resolved by processing a conventional HSLA steel, usually one containing vanadium and often with an enhanced nitrogen content of 0 015 wt % [Davies, 1978, Owen, 1980], in such a way as to change the microstructure from one of ferrite-pearlite to one of ferrite containing islands of martensite-austenite to introduce continuous yielding and a high work hardening rate However, there are problems concerning spring back and control of the processing treatment

2 2 8 Inclusion Shape Control

The elongation of non-metallic inclusion, which occurs during rolling of steel, is responsible for the anisotropy in ductility and toughness The effects of elongated

co-planar arrays of manganese sulfide are generally considered most deleterious, because they lead to lower ductility and toughness values in the transverse direction Around 1965 it was realized that these elongated or co-planar arrays are detrimental to longitudinal bending of hot rolled strips [Korchynsky and Stuart, 1970] and caused lamellar tearing in highly restrained welds in thick plates [Farrar, 1971] The problem has, however, been overcome by the important development of inclusion shape control technology in which the plasticity of MnS inclusions is decreased by the addition of zirconium [Michelich et al , 1971], cerium [Little and Henderson, 1971], or calcium [Hilty and Popp, 1969], so that elongated stringers are not produced

2 3 Classifications

HSLA steels can be divided into the following six categories

- Weathering steels, which contain small amounts of copper and phosphorus, and display improved atmospheric corrosion resistance and solid solution strengthening
- Microalloyed ferrite-pearlite steels, which contain very small additions of strong carbide or carbonitride forming elements such as niobium, vanadium and/or titanium for precipitation strengthening, grain refinement and transformation control
- As-rolled pearlitic steels, which may include corbon-manganese steel but which also have small addition of other alloying addition to enhance strength, toughness, formability and weldability
- Acicular ferrite steels, which are low carbon steels with an excellent combination of high yield strength, formability, weldability and toughness

- Dual-phase steels, which have a microstructure of martensite dispersed in a ferrite matrix and provide a good combination of ductility and high tensile strength
- Inclusion shape controlled steels, which provide improved ductility and toughness by small addition of calcium, zirconium or titanium, or rare-earth elements so that the shop of sulfide inclusion are changed from elongated stringers to small, dispersed, almost spherical globules

2 3 1 Microalloyed Ferrite-Pearlite Steels

The yield strength σ_y, is given in general terms by

$$\sigma_y = \sigma_i + \sigma_s + \sigma_p + \sigma_d + k_y d^{1/2}$$

where σ_i is the friction stress opposing dislocation motion, σ_s is the contribution from solid solution strengthening, σ_p is due to precipitation strengthening, σ_d is because of dislocation strengthening, k_y is the dislocation locking term and d is the ferrite grain diameter Using well known theories of strengthening and work hardening of structures comprising unreformed particles in an unreformed matrix [Ashby, 1966, Brown and Stobbs, 1971] it has been shown that the flow stress and the work hardening rate can be predicted accurately [Ballinger and Gladman, 1981, Lanzillotto and Pickering, 1982] The flow stress and the work hardening rate are controlled by the volume fraction and the particle size of the M-A constituent Theories predict that both the flow stress and the work hardening rate should be related linearly to the reciprocal of the square root of the size of the M-A particles

These steels use additions of niobium and vanadium to increase the strength of hot rolled steel without an increase in the carbon and/or manganese contents The various types of microalloyed ferrite-pearlite steels include

- Vanadium-microalloyed steels
- Niobium-microalloyed steels
- Niobium-molybdenum steels
- Vanadium-niobium microalloyed steels
- Vanadium-nitrogen microalloyed steels
- Titanium-microalloyed steels
- Niobium-titanium microalloyed steels
- Vanadium-titanium microalloyed steels

Vanadium containing steels are used in the hot-rolled condition and also in the controlled-rolled, normalized, or quenched and tempered condition Strengthening from vanadium averages between 5 and 15 MPa per 0 01 wt % V [Porter, 1986] depending on carbon content and cooling rate An optimal level of precipitation strengthening occurs at a cooling rate of about 170C min-1 Manganese content and ferrite grain size also affect the strengthening of vanadium microalloyed steels [Meyer et al , 1977] The effect of manganese on precipitation strengthening is more in vanadium steels than in niobium steels

2 4 Role of Microalloying Elements

Alloying elements are selected to influence the transformation temperatures so that the transformation of austenite to ferrite and pearlite occurs at a lower temperature during air-cooling This lowering of the transformation temperature produces a finer grain transformation product, which is the main source of strengthening At the low carbon levels typical of HSLA steels, elements such as silicon, copper, nickel and phosphorous air particular effective for producing fine pearlite

The microalloying elements Nb, V and Ti are added singly or in combination to what are essentially carbon-manganese steels, to produce the HSLA steels These three microalloying elements have very different effects due to their different affinities for carbon and nitrogen Microalloying elements are added to steel for two main purposes, namely, to grain refine and/or precipitation strengthening Both effects result from the precipitation of microalloy carbides, nitrides or carbonitrides These precipitates in ferrite can prevent ferrite grain growth during or after the transformation and so have an indirect grain refining effect It must be emphasized, however, that microalloy carbides/nitrides precipitated in the austenite do not cause strengthening

2 4 1 Solubility Effects

The solubility of the microalloy carbides/nitrides in austenite decreases in the following order TiN, NbN, TiC, VN, NbC, VC and the same order is followed for the solubility in ferrite in which the solubility is some two orders of magnitude smaller than in austenite A general solubility product equation is

$$\log[M][X]=A/T+B$$

where [M] and [X] are the weight percent of microalloying element and C or N dissolved in the austenite at T Kelvin, and A and B are constants

2 4 2 Grain Refinement

The essence of grain refinement is to prevent grain boundary migration and grain growth This may be done either by grain boundary solute segregation introducing a frictional drag on the moving boundary or by grain boundary pinning particles which decrease the grain boundary area and hence the overall grain boundary energy

If the grain size is to remain small at the high temperature in the austenite region, such as what is used for rehearing for hot rolling and especially for re crystallization controlled rolling (RCR), two conditions are to be satisfied [Gladman and Pickering, 1967]

(i) the volume fraction of pinning particles must remain large, and

(ii) the pinning particles must grow only very slowly at the temperature involved

The low solubility and general thermodynamic stability causes TiN to be the most resistant of the microalloying carbides/nitrides to particle coarsening, and thus the most effective grain boundary pinning phase This partially explains why Ti microalloying is used in recrystallization controlled rolled steels

2 4 3 Precipitation Strengthening

In Nb steels the strengthening precipitates are predominantly NbC and in Ti steels TiC Due to the restricted solubility of Ti in the austenite, strengthening by TiC requires rather high Ti additions compared to the 0 01/0 02 wt % required for optimal grain refinement A higher austenizing temperature will also be required Thus, Ti additions are not predominantly used for precipitation strengthening in may steels In contrast, due to the much increased solubility of VC, V additions are mainly made for precipitation strengthening and unlike NbC, the solubility and hence the precipitation potential is not greatly limited by the carbon content, eight due to additional VN precipitation or N being dissolved in VC as V(CN) Also, decreasing the transformation temperature produces finer precipitates, and greater precipitation strengthening

2 4 4 Alloying Elements

Vanadium Precipitation strengthening is one of the primary contributors to strength in microalloyed steels, it is most readily achieved with vanadium additions in the 0 03 to 0 10 wt % range Vanadium also increases toughness by stabilizing dissolved nitrogen The impact transition temperature also increases when vanadium is added

Niobium Niobium can also have a strong precipitation strengthening effect provided it is taken into solution during reheat and is kept in solution during forging Its main contributions, however, are to form precipitates above the transformation temperature and to retard the recrystallization of austenite, thus promoting a fine-grained microstructure having improved strength and toughness Concentrations vary from 0 020 to 0 10 wt % Nb

Titanium Titanium can behave both as a grain refiner and precipitation strengthener, depending on its content At compositions greater that 0 050 wt %, titanium carbides begin to exert a strengthening effect However, at this time, titanium is used commercially to retard austenite grain growth and thus improve toughness Typically titanium concentrations range from 0 01 to 0 020 wt %

Molybdenum Molybdenum in hot rolled HSLA steels is used primarily to improve hardenability when transformation products other than ferrite or pearlite is desired It also greatly simplifies the process controls necessary in the forge shop

Manganese Manganese is generally present in larger quantity in HSLA steels that in structural carbon steels It functions mainly as a mild solid solution strengthener in ferrite and also provides a lowering of the austenite to ferrite transformation temperature In addition, it improves the notch toughness of HSLA

steels In steels for weldıng applıcatıon, Mn should be kept below some maxımum value that depends on the overall composıtıon but maınly on the carbon content

Sılıcon Sılıcon ıs usually present ın fully deoxıdızed steels ın amounts up to 0 35% Sılıcon has a strengthenıng effect ın low alloy structural steels In larger amounts ıt reduces scalıng at elevated temperature Sılıcon has a sıgnıficant effect on yıeld strength enhancement by solıd solutıon strengthenıng Effects of selected elements on the varıous factors dıscussed above are gıven ın Table 2 2

Table 2 2 Effect of selected elements on the propertıes of mıcroalloyed steels[Wrıght, 1990]

Elements	**Pı ecıpıtatıon Sterengthenıng**	**Feı rıte graın refinement**	**Nıtrogen fixıng**	**Structure modıficatıon**
Vanadıum	Strong	Weak	Strong	Moderate
Nıobıum	Moderate	Strong	Weak	None
molybdenum	Weak	None	None	Strong
Tıtanıum	Strong(< 05%Tı	Strong	Strong	None

2 5 Steel-Makıng

In general, from the poınt of vıew of steel makıng HSLA steels are not dıfficult to produce Nıobıum does not have a strong affinıty for eıther nıtrogen or oxygen ın lıquıd steel, and therefore can be added to steels whıch show a wıde range of levels of deoxıdatıon and solıdıficatıon rates

Vanadıum, lıke nıobıum, has no strong deoxıdızıng effect ın lıquıd steel but, due to greater solubılıty of ıts carbıdes/nıtrıdes compared wıth those of Nb, ıt ıs less

susceptible to strain induces precipitation and therefore produces less severe loss of hot ductility than does Nb

Titanium has a strong affinity for O, N and S and also for C, and therefore poor and variable recoveries of Ti are experienced unless the steel is quite heavily deoxidized, for example by Al Inert gas shrouding of liquid metal is required to prevent oxidation of Ti by air

2 6 Thermomechanical Treatments for Sheets, Strips and Plates

The aim of processing of HSLA steel is to condition the austenite that it transforms to produce the finest ferrite grain size in order to achieve the greatest yield strength consistent with optimal toughness and ductility

2 6 1 Thermomechanical Working and Controlled Rolling

It is required that in order to produce the finest ferrite grain size, the thermomechanically worked austenite should, during transformation, exhibit a high ferrite nucleation rate and a low ferrite growth rate In addition, the nucleation and growth of microalloy carbides/nitrides are of utmost importance

The requirements to produce the necessary fine ferrite grain size are greatest area of austenite grain boundary for ferrite nucleation Such nucleation can occur on deformation bands in the unrecrystallized austenite, on the recovered sub-structure boundaries, particularly if these contain precipitates, and on undissolved carbide/nitride particles [Amin and Pickering, 1982]

In order to obtain the finest possible ferrite grain size, not only the initial austenite grain size should be as fine as possible but also the temperature of rolling

should be as low as possible provided it is above the recrystallization stop temperature, and the rolling strain should be as large as possible

Current steels tend to employ combinations, e g Nb-V The NbC or VN (using enhanced nitrogen) tend to restrict grain growth, whilst the more soluble VC is used to precipitation strengthen the ferrite Microalloying additions also have another important effect during controlled rolling, in that they retard recrystallization The effect seems to be in the ascending order of effectiveness Mn, Al, V, Nb and Ti on an atomic percentage basis is well established that the recrystallization stop temperature of the austenite increases in the increasing order V, Al, Ti, Nb The importance of retardation on recrystallization during controlled rolling lies in the ability to use a low finish rolling temperature to produce elongated unrecrystallized austenite grains, which can transform to very fine polygonal ferrite [Tanaka et al , 1977]

2 6 1 1 Conventional Low Temperature Controlled Rolling

Tanaka (1981) identified the processing parameters to condition the austenite to produce the finest ferrite grain size and optimum precipitation strengthening, using conventional low temperature controlled rolling as

(a) a low reheating temperature to produce a fine initial austenite grain size,

(b) austenite grain refinement by recrystallization,

(c) suitable pass schedules and reductions to obtain in the initial roughing passes a fine, uniform recrystallized austenite,

(d) delay between roughing in the recrystallization regime and finishing in the unrecrystallized regime,

(e) suitable reductions in the unrecrystallized regime below the recrystallization stop temperature, and in some cases finishing below the Ar3

2 6 1 2 Recrystallization Controlled Rolling

Much work has been done in recent years on recrystallization controlled rolling [Roberts, 1984, Roberts et al , 1984, Zheng et al , 1984, Fix et al , 1986, Korchynsky, 1990] An initial very fine austenite grain size is required prior to rolling and this is achieved by addition of Ti to form TiN Such particles very effectively pin the austenite grain boundaries and inhibit grain growth Rolling is then carried out above the recrystallization stop temperature so that repeated recrystallization occurs during deformation sequences The finest recrystallized austenite grain sizes are produced by heavy rolling reductions at the lowest possible temperature above the recrystallization stop temperature The fine recrystallization austenite grains have a high grain boundary surface area per unit volume, i e a high Sv, and thus provide many ferrite nucleation sites Hence, on transforming the austenite, a very fine ferrite grain size is obtained which can be 10-m or less and is produced from a recrystallized austenite 20-m or less [Zajac et al , 1990, Korchynsky, 1990] Using a 0 09 wt % C, 1 4 wt % Mn, 0 01 wt % Ti, 0 08 wt % V, 0 013 wt % N steel, recrystallization controlled rolled and cooled at 9oC s-1, one can obtain a yield stress of 450/500 MPa with as ITT of-70oC [Zajac et al, 1990] which is comparable with the properties produced by conventional low temperature controlled rolling

2.6.2 Controlled Cooling

The purpose of controlled or accelerated cooling is to produce the optimal γ to α transformation temperature so that the finest ferrite grain size is achieved, together with optimal strengthening by microalloy carbide/nitride precipitation A useful method of refining the ferrite grain size for a given austenite grain size is decrease the

transformation temperature This increases the ferrite nucleation rate and the effect may be achieved by alloying or by increasing the cooling rate The greater the cooling rate or lower the coiling temperature, the finer is the ferrite grain size from given austenite structure

Using the well known solubility relationships to assess the volume fraction precipitates, and the observed effects of coiling temperature on both the ferrite grain size and the precipitate size, it is possible by employing a Petch type equation and an Ashby Orowan precipitation strengthening equation, to calculate the yield stress as a function or coiling temperature This can be converted into a nomogram for process control

2 7 HSLA Forgings

Heat treated forgings currently outperform alternative materials where strength, toughness and reliability are primary considerations However, methods must be found that achieve these benefits at lower cost

Unfortunately, the thermomechanical processing used for HSLA flat rolled products cannot be readily transformed to hot forgings Typical forgings are produced from bars that are induction-heated or gas heated to 1250°C, then hot worked to 1100°C Grain growth and precipitate coarsening are rapid at these temperatures Attempts to forge at a lower temperature to optimize the as-forged steel properties result in increased die wear and decreased equipment efficiency

2 7 1 Metallurgical Effects

Upon cooling from forging temperature, niobium carbide begins to precipitate at about 1205 °C The precipitate continues to form and coarsen as temperature falls to

925 ^{0}C the absence of hot workıng Contınued hot workıng ınto the 900 ^{0}C temperature rang however, retards austenıte recrystallızatıon and precıpıtatıon resultıng ın the development of a refined austenıte graın sıze

Vanadıum carbonıtrıde precıpıtatıon begıns at about 950oC Precıpıtate coarsenıng durıng accelerated coolıng from the forge temperature ıs mınımal, maxımum precıpıtatıon strengthenıng effect ıs achıeved Toughness ımproves when ferrıte graın sıze ıs mınımızed by controllıng the finısh hot-workıng temperature The 20J-transıtıon temperature for a nıobıum contaınıng steel ıs reduced from 30 °C to –80 °C by reducıng the finısh hot-woıkıng temperature from 1050 to 900oC The ımprovement ıs ındependent of reheat temperature [Metal Handbook, 1990]

2 7 2 Development of HSLA Forgıng Steels

A West German composıtıon 49MnVs3 was successfully used for automotıve connectıng rods The steel was typıcal of the first generatıon of mıcroally steels wıth a medıum carbon content and addıtıonal strengthenıng through vanadıum carbonıtrıde precıpıtatıon The parts were subjected to accelerated aır coolıng dırectly from the forgıng temperature

(a) Fırst Generatıon Mıcroalloyed Forgıng Steels

These steels generally have ferrıte-pearlıte mıcrostructures, tensıle strength above 760 MPa and yıeld strength ın excess of 450 MPa The room temperature Charpy V-notch toughness of first generatıon forgıngs ıs typıcally 7-14 J It became apparent the toughness would have to be sıgnıficantly ımproved to realıze the full potentıal mıcroalloy steel forgıngs

(b) Second Generatıon Mıcroalloyed Forgıng Steels

Second generatıon mıcroalloyed forgıng steels were ıntorduced ın about 1986 These are typıfied by the West German grade 26MnSıVS7 The carbon content of these steels was reduced to between 0 1 to 0 3% They are produced wıth eıther a ferrıte-pearlıte mıcrostructure or ın acıcular ferrıte structure The latter results from the suppressıon of the pearlıte transformatıon caused by the addıtıon of about 0 1% Mo Tıtanıum addıtıon ıs also made to these steels to ımprove ımpact toughness further One of the prımary concern of any steel user ıs the consıstency of the fınıshed part propertıes Heat treatment has successfully addressed thıs concern, and a method must be found to ensure consıstency of propertıes ın fınıshed parts made from mıcroalloyed steels One dısadvantage of the ferrıte-pearlıte mıcroalloyed steel ıs that the fınıshed strength and hardness are a functıon of coolıng rate Coolıng rate can vary because of process changes or part geometry The molybdenum treated steel removes the varıable of accelerated coolıng from the process, an advantage to the hot forger and end user because conveyors are unnecessary

(c) Thırd Generatıon Mıcroalloyed Forgıng Steels

Thırd generatıon mıcroalloyed forgıng steels went ınto commercıal productıon ın the Unıted States ın 1989 These steels dıffer from theır predecessors ın that they are dırectly quenched from the forgıng temperature to mıcrostructures of lath martensıte wıth unıformly dıstrıbuted temper carbıdes

The newest generatıon of mıcroalloyed forgıng steels has five to sıx tımes the toughness at -30oC and twıce the yıeld strength of the second-generatıon materıals No specıal forgıng practıces are requıred except for the use of a water coolıng system In a comparıson of thırd generatıon mıcroalloyed steels wıth two standard quenched and

tempered carbon and alloy steels (1040 and 4140), the properties of the direct quenched microalloyed grade were very similar to those of quenched and tempered 4140 The hardness of all the three materials tested was HRC

2 8 Mechanical Properties of HSLA Steels

Blarasin and Farsetti (1989) tested five microalloyed steels containing V, Nb and/or Ni in the carbon range of 0 21 to 0 41 wt % for tensile and impact properties Yield strength varied from 590 to 840 MPa and UTS from 800 to 940 MPa Room temperature impact energy was reported to lie in the range of 80 to 140 J cm-2 AISI 1141 vanadium microalloyed medium carbon steel in the as - forged condition has a yield strength (0 2% offset) of 524 MPa and tensile strength of 875 MPa [Yang and Fatemi, 1995] Low carbon V/Nb microalloyed steels have been reported to have slightly higher compressive yield stress than the tensile value [Kim and Fine, 1982] Compressive yield strength for the four steels studied was found to lie in the range of 450-610 MPa HSLA steels also have lower ductile to brittle transition temperature Typically a Charpy V-notch fracture appearance transition temperature of -40oC is achieved Conventional controlled rolling can give excellent combination of strength (yield strength 450-525 MPa) and toughness (ITT as low as - 80oC) ITT of low carbon ferrite-pearlite microstructure is given by Glandman and Pickering, (1983) as

$$ITT\ (^{0}C = -19 + 44^{*}\ (\text{wt}\ \%\ Si) + 700^{*}\ (\text{wt}\ \%\ N_{free})^{1/2} + 2\ 2\ (\%\ \text{pearlite}) - 11\ 5\ d^{-1/2}$$

CHAPTER 3

EXPERIMENTAL PROCEDURE

3 1 MATERIAL FOR THE PRESENT INVESTIGATION

A medium carbon microalloyed forging steel 49MnVS3 was chosen for this investigation The above steel finds application in the automotive industry The microalloyed steel 49MnVS3 is used for making crankshafts and connecting rods, which are subjected to fluctuating loads The material used in this investigation was produced at Tata Iron and Steel Company, Jamshedpur, India

Chemical composition of the supplied material has been shown in Table 3 1

Table 3 1 Chemical composition of the Material

Elements	C	Si	Mn	P	S	Al	V	N	Fe
(Wt %)	0 49	0 27	0 98	0 02	0 05	0 02	0 09	70ppm	Bal

3 2 THERMO-MECHANICAL TREATMENT

Thermo-mechanical processing in the present investigation was done by hot rolling of the alloy by imparting the known amount of thickness deformation at a given temperature either in the Austenite or the (Austenite+Ferrite) phese field followed by water quenching or Two-Step-Cooling

3 2 1 Equipments for Thermo-Mechanical Processing

Heating and soaking of the samples was done in a specially designed high temperature furnace, kept very close to the rolling mill Muffle of the furnace consisted of an inconel tube and was closed from one end Argon gas was introduced into the furnace

through a 4 mm internal diameter stainless steel tube passing through the closed end of the chamber Thus, the furnace was capable of maintaining a protective atmosphere during heating and soaking of the alloy Further, the furnace was mounted on wheels so as to bring it very close to the rolling mill The furnace was heated by silicon carbide rods and had a constant temperature zone of length of about 17 5 cm To protect the steel from oxygen absorption during its heating and soaking grade I pure Ar gas was continuously passed through the furnace

Hot rolling of flat specimen of steel was done on a 2-high rolling mill having rolls of 135 mm diameter Speed of rotation for the rolling mill was kept constant at 55 rpm in all the experiments No prior heating of the rolls was done before hot rolling of the specimens and they were maintained at room temperature Prior to the thermal treatment, the samples were placed on a small, perforated Inconel tray fitted with a long handle and then pushed carefully into the constant temperature zone of the furnace

Hot rolling was essentially multi-pass in nature To nullify the temperature drop effect a soaking treatment of two minutes was given in between each pass Before hot rolling the samples were taken out from the furnace and were quickly fed (within 3-4 seconds) to the feeding end of the rolling mill Thus, there was a very little temperature drop during rolling The samples were quickly quenched into water after the last pass of the given hot rolling schedule

3 2.2 working schedule

The steel samples were rolled at 760°C, 800°C, 900°C, 1000°C, 1100°C and 1200°C These samples, prior to their hot rolling were soaked under pure argon atmosphere for time period ranging from 1 hour to 20 minutes (details are given in figure 3 1 and 3 2)

All the samples had an initial thickness of ~ 7 mm and they were given ~ 50% thickness reduction in four passes In between each pass the samples were put back into the furnace and were soaked for ~ 2 minutes before carrying out the next stage of the hot rolling After hot rolling at each temperature a set of sample were directly quenched in water On the other hand in two stage cooling process other set of sample were cooled in still air up to 660°C followed by water quenching In addition, water quenched samples were subjected to further tempering at 560°C for 1hr and TSC samples were annealed at 420°C for 1 hr

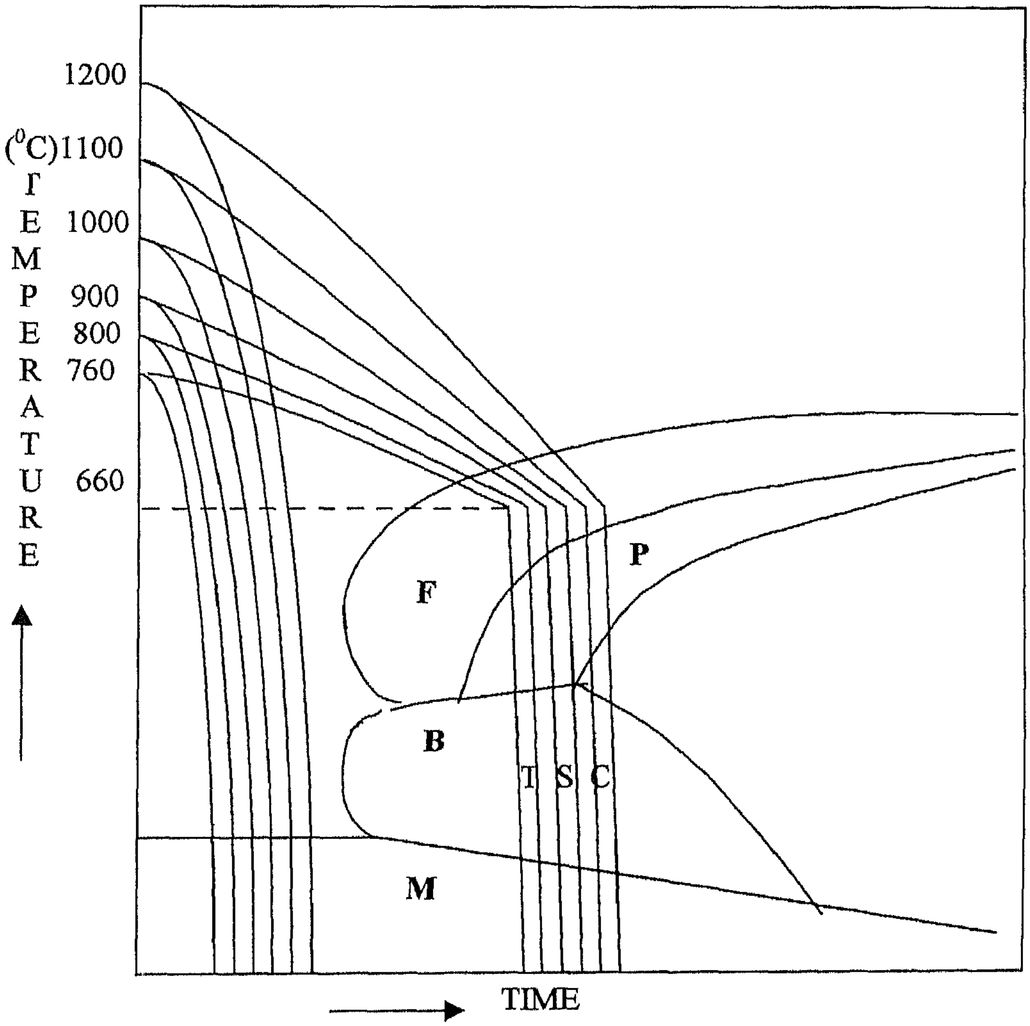


Fig 3 1 Schematic Representation of Thermomechanical processes Studies

Quench and Tempered Route

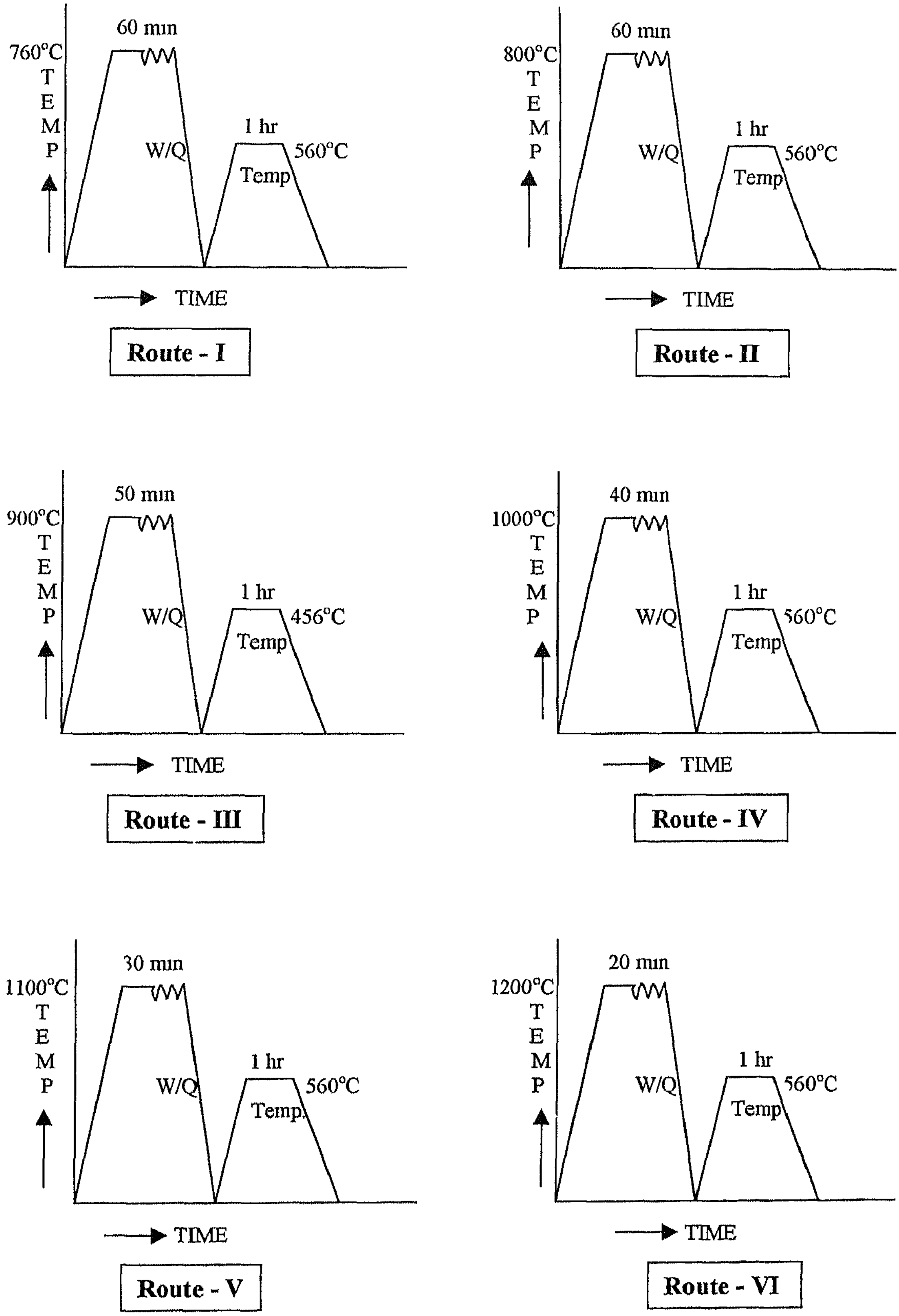


Figure 3 2 Schematic representation of thermomechanical treatment (quench and tempered route) given to 49MnVS3

Two-Step cooled Route

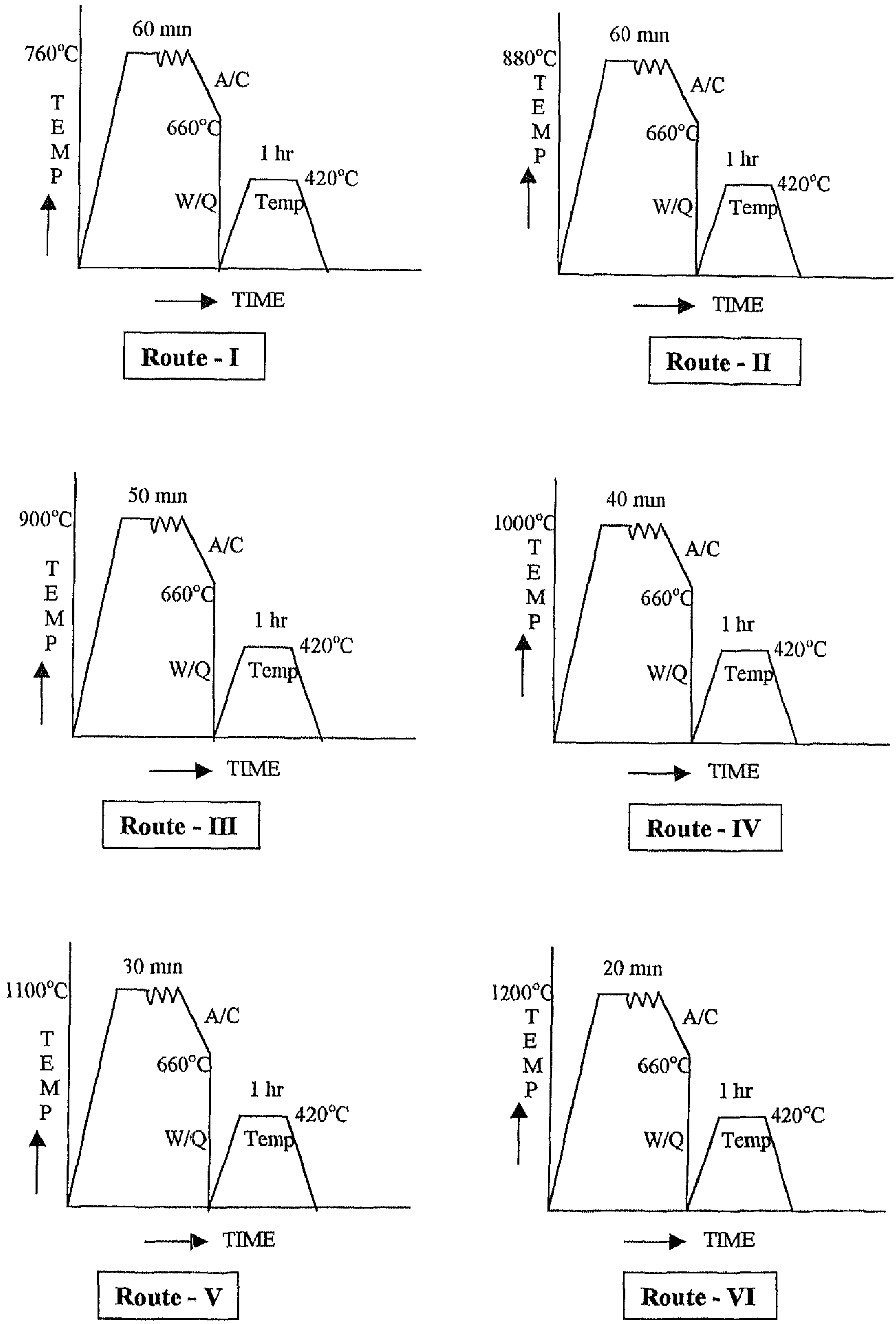


Figure 3 3 Schematic representation of thermomechanical treatment (two-stap cooled route) given to 49MnVS3

3 3 Microstructural Characterization

Microstructures taken from all the specimens were examined by optical and scanning electron microscopes Prior to the observations, all the samples were mechanically polished on emery papers (0 to 4 grades) followed by wheel polishing with alumina powders The samples were then etched with 2% Nital reagent (2%HNO_3 +98% Mithanal) Microstructural observations were, generally, done on transverse cross sections of the specimens, unless otherwise mentioned

Optica microscopy was done under a Leitz Wetzlar optical microscope

Scanning electron microscopy was done under a JEOL-JSM 840A Scanning electron microscope The samples were observed under the scanning electron microscope, operated at 10 kV using secondary electron radiation (SE mode)

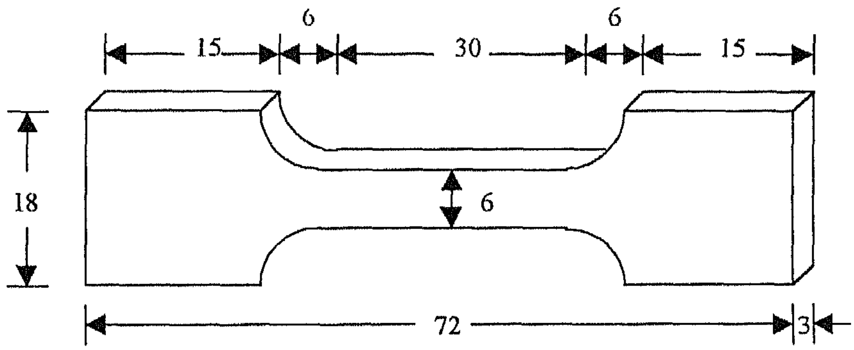
3 4 Mechanical Testing

3 4 1 Hardness Measurements

The hardness of various phases was measured using a Leitz mini-load micro hardness tester Microhardness was measured using the etched samples for microscopy Microhardness was measured using 0 49N load Microhardness measurements were done in 5-6 different points for reproducibility Average of both the diagonals of the indentation was converted to the corresponding HV

3 4 2 Tensile Testing

Tensile specimen was prepared according to ASTM standard Tensile tests were done at a strain rate of $5*10^{4}$ in strain control mode using a closed loop servo–hydraulic material testing machine The dimensions of the tensile specimen are given in figure 3 2



All dimensions are in mm

Fig 3 2 The dimensions of the tensile specimen used in the present investigation

CHAPTER 4

RESULT AND DISCUSSIONS

As mentioned in Chapter 3, the steel samples were rolled at 760°C, 800°C, 900°C, 1000°C, 1100°C and 1200°C These samples, prior to their hot rolling were soaked under pure argon atmosphere for time period ranging from 1 hour to 20 minutes (details are given in Chapter 3) All the samples had an initial thickness of ~ 7 mm and they were given ~ 50% thickness reduction in four passes In between each pass the samples were put back into the furnace and were soaked for ~ 2 minutes before carrying out the next stage of the hot rolling Subsequently the samples were either quenched in water or were cooled through two stage cooling (TSC) In addition, a set of samples was subjected to further tempering/annealing after quenching/TSC The microstructures of thermomechanically treated 49MnVS3 microalloyed steel samples thus obtained were characterized by various techniques already discussed in Chapter 3 Similarly, the mechanical properties of each of thermomechanically treated steel were measured Results obtained on them have been described and discussed in the present Chapter

4 1 Structure of the Steel prior to Thermomechanical Treatment:

As explained in Chapter 3, the 49MnVS3 steel was received in the as-forged condition and was taken for various thermomechanical treatments The as-forged microstructure of the steel, as given in Fig 4 1, shows a typical pearlite-ferrite structure Volume fractions of pearlite and ferrite were estimated through quatitative metallographic technique and were found to be 66% and 34% respectively The pearlite colony size and inter-lamella spacing, as measured in an earlier investigation, were found to be 16 ± 2 2 μm and 0 25 ± 0 03 μm respectively [Sarma, 1998]

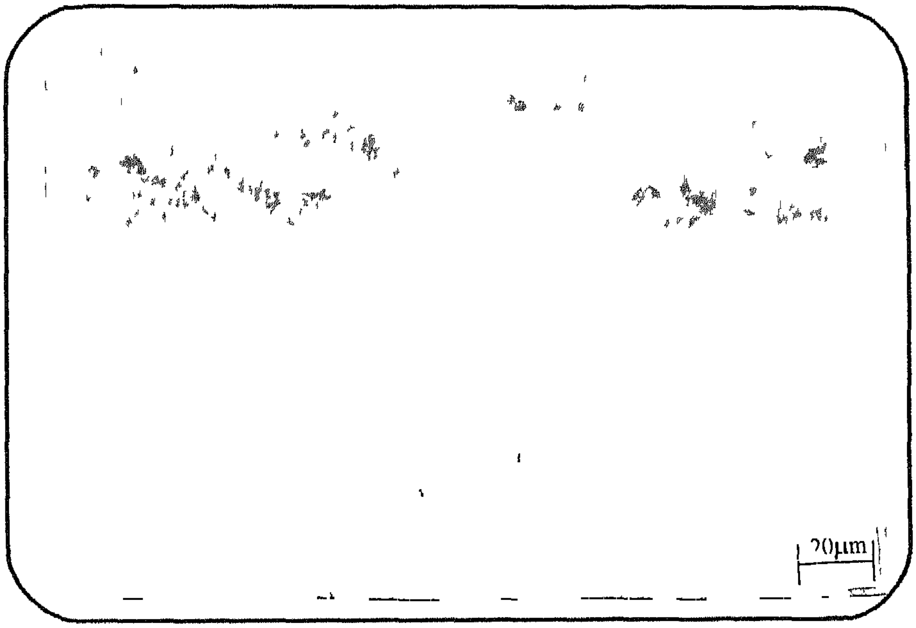


Figure 4 1 Microstructure of As-Received Material

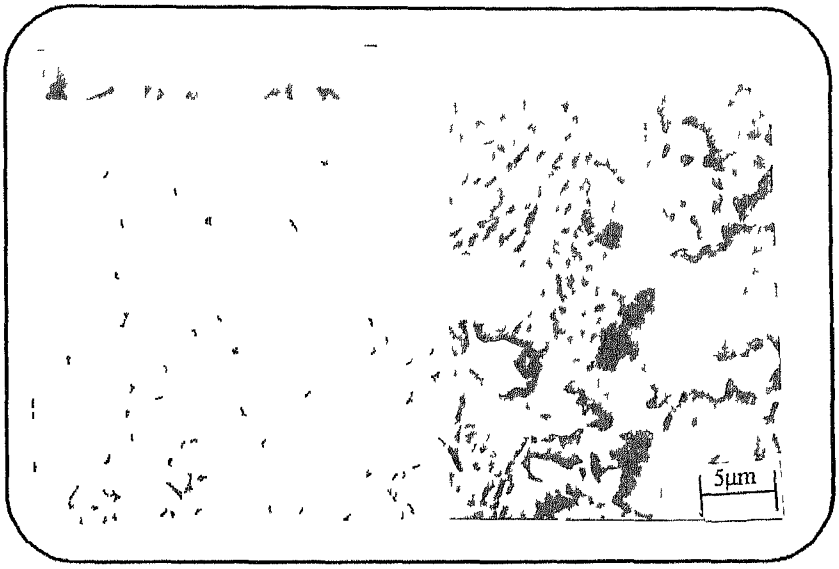


Figure 4 2 Microstructure of the sample rolled at 760^{0}C followed by water quenching

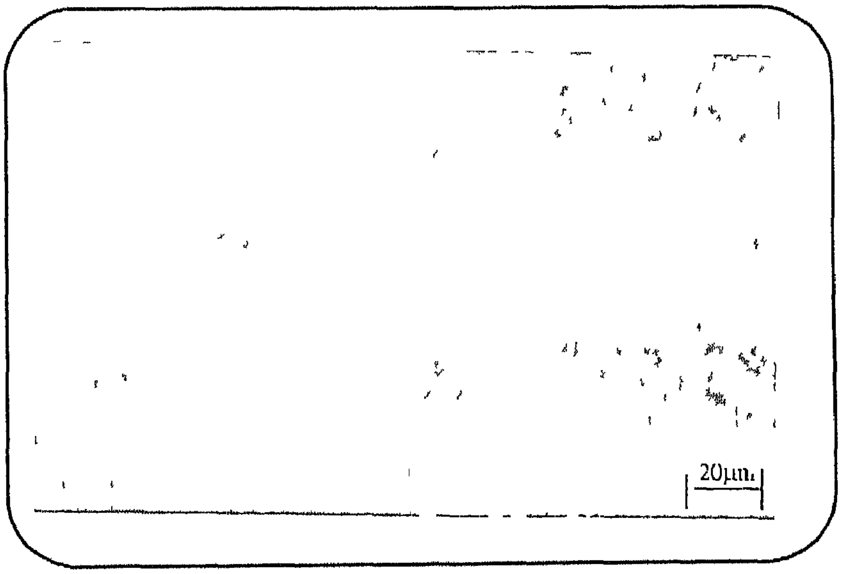


Figure 4 3 Microstructure of sample rolled at 800°C followed by water quenching

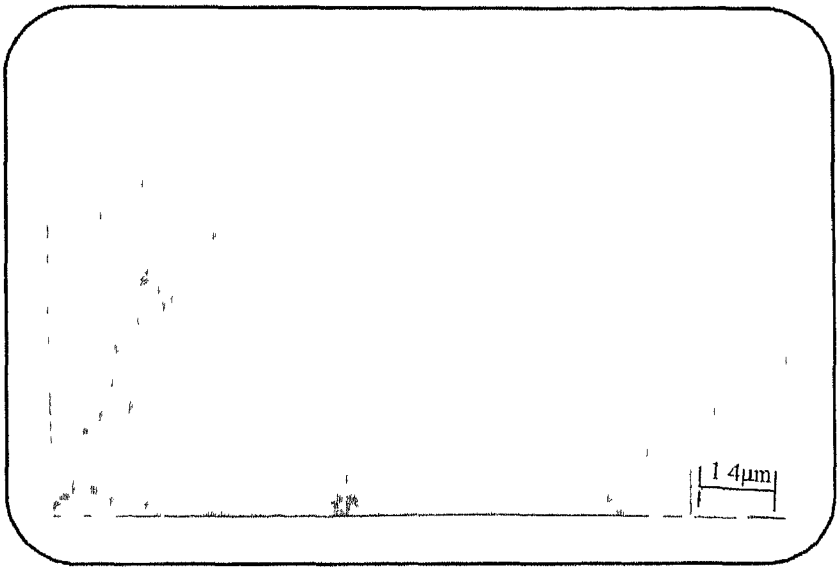


Figure 4 4 Microstructure of sample rolled at 900°C followed by water quenching

Figure 4 5 Microstructure of sample rolled at 900°C followed by water quenching and annealed at 560°C for 1hr

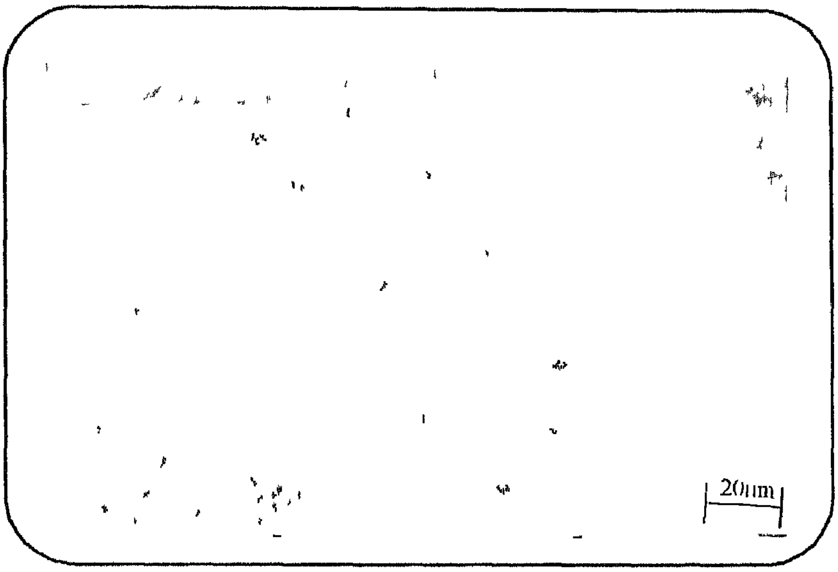


Figure 4 6 Microstructure of sample rolled at 1000°C followed by water quenching

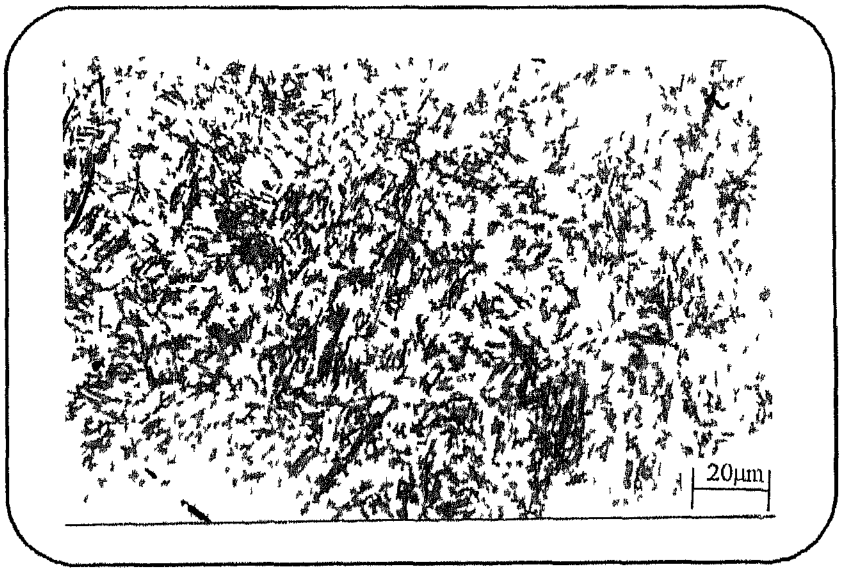


Figure 4 7 Microstructure of sample rolled at 1100°C followed by water quenching

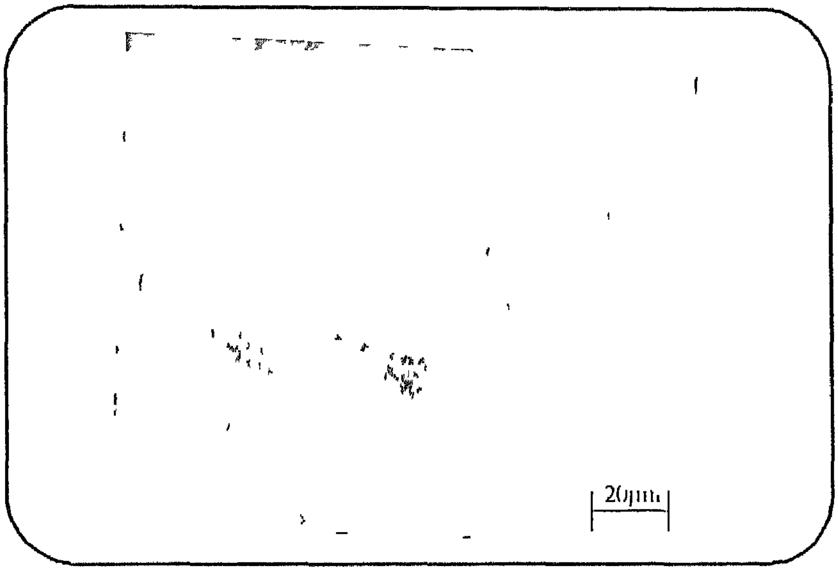


Figure 4 8 Microstructure of sample rolled at 1200°C followed by water quenching and annealed at 560°C for 1hr

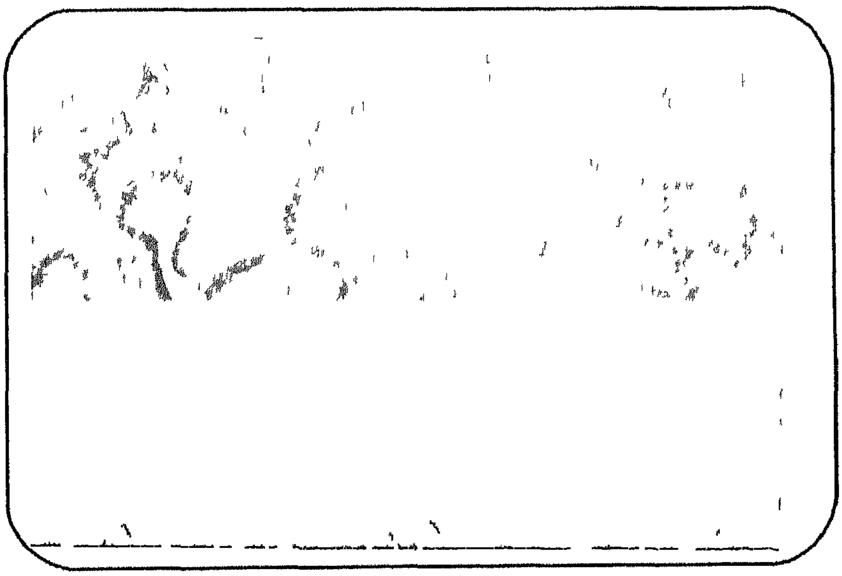


Figure 4 9 Microstructure of the sample rolled at 760°C followed by two-step cooling(TSC)

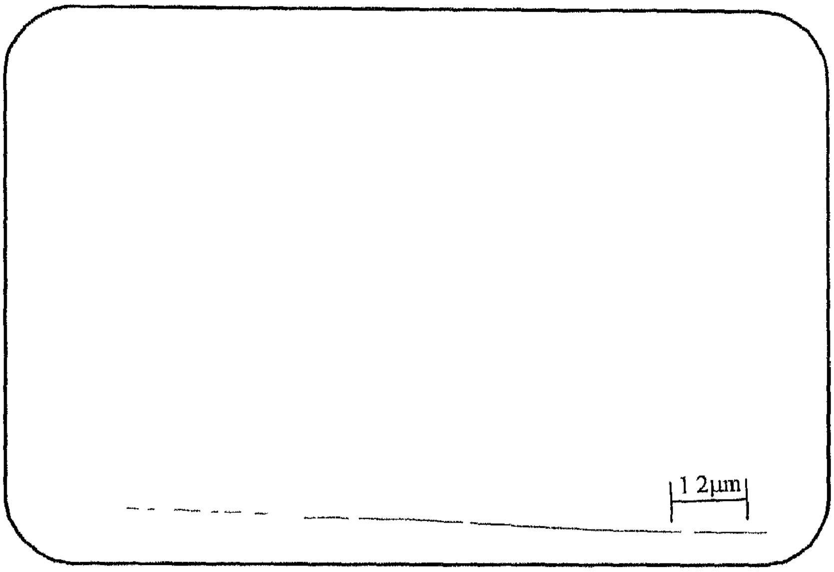


Figure 4 10 Microstructure of the sample rolled at 800°C followed by TSC

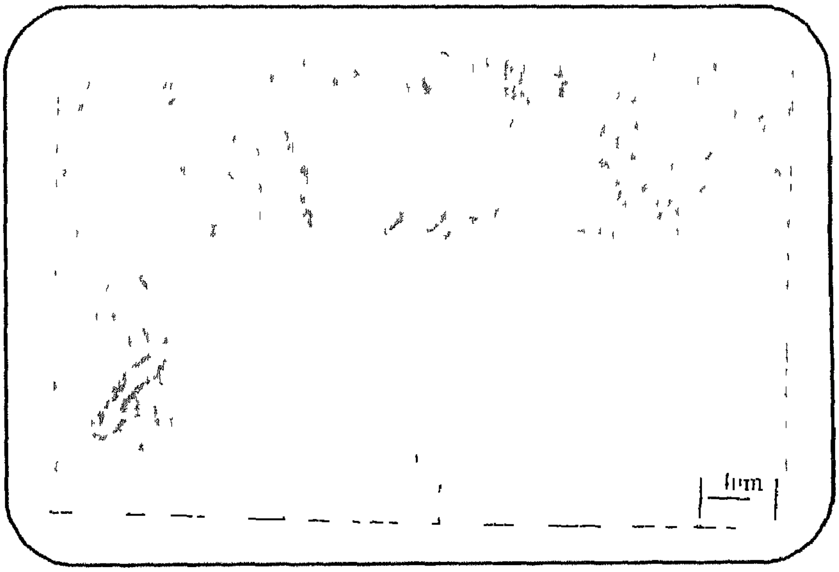


Figure 4 11 Microstructure of the sample rolled at 900°C followed by TSC

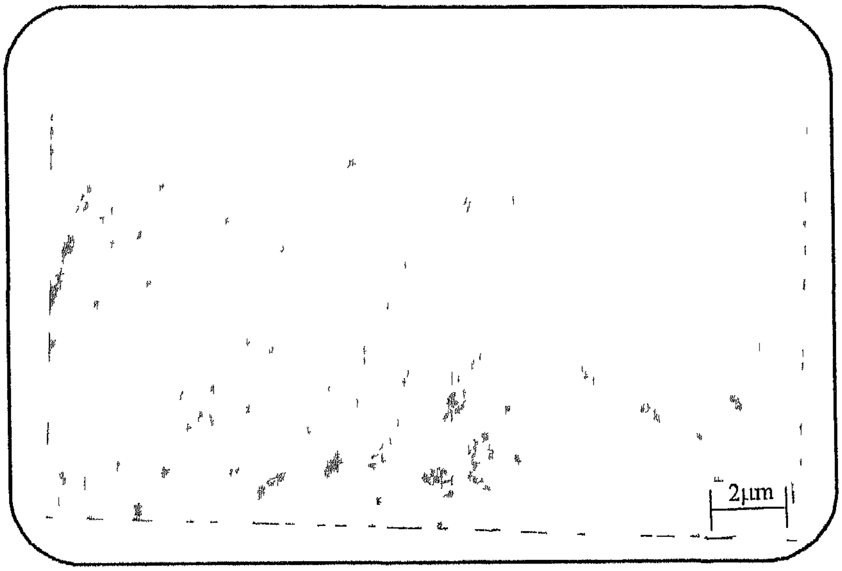


Figure 4 12 Microstructure of the sample rolled at 900°C followed by TSC

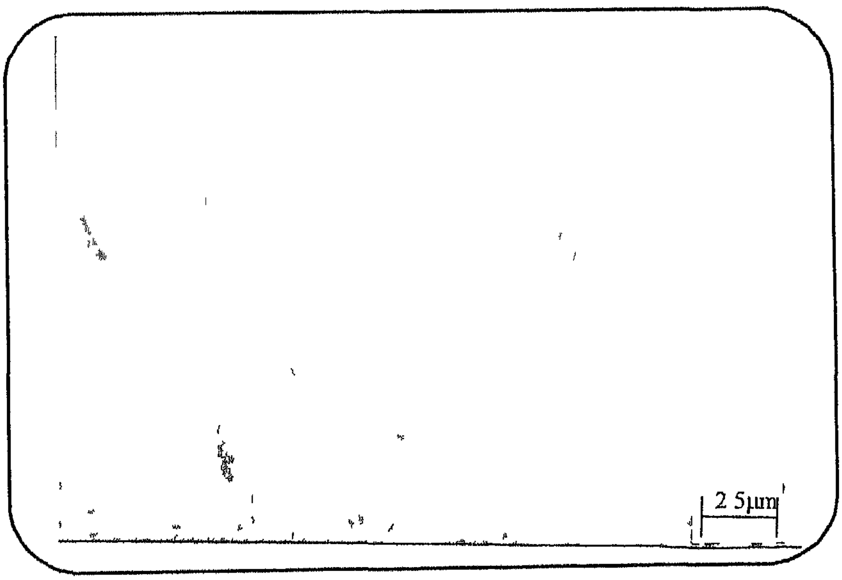


Figure 4 13 Microstructure of the sample rolled at 1000°C followed by TSC

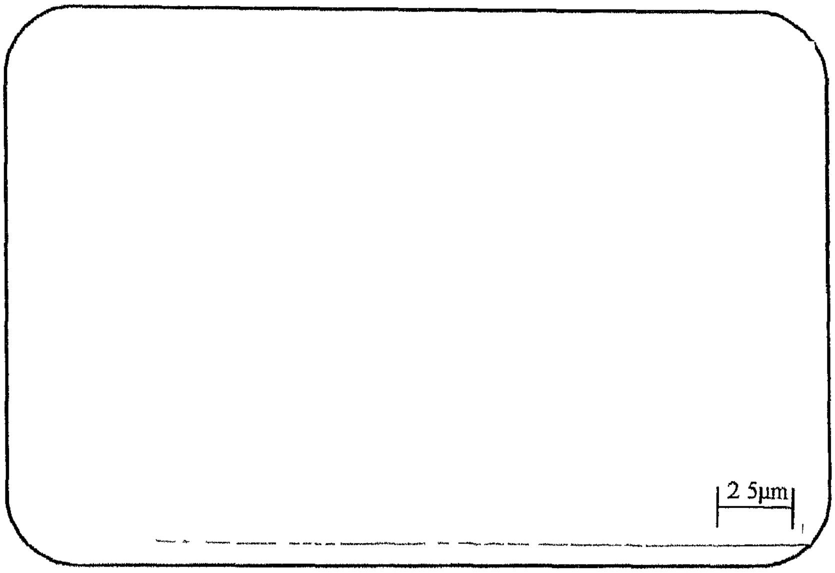


Figure 4 14 Microstructure of the sample rolled at 1000°C followed by TSC and annealed at 420°C for 1hr

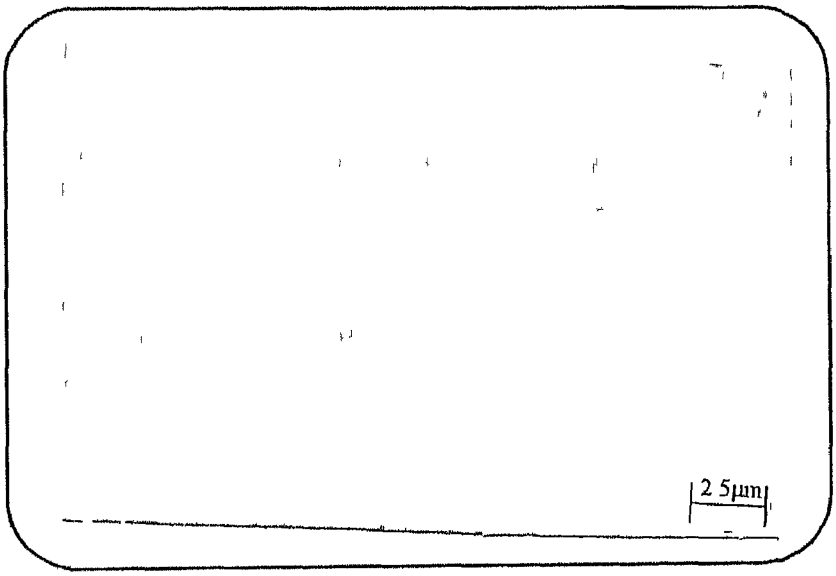


Figure 4 15 Microstructure of the sample rolled at 1000°C followed by TSC and annealed at 420°C for 1hr

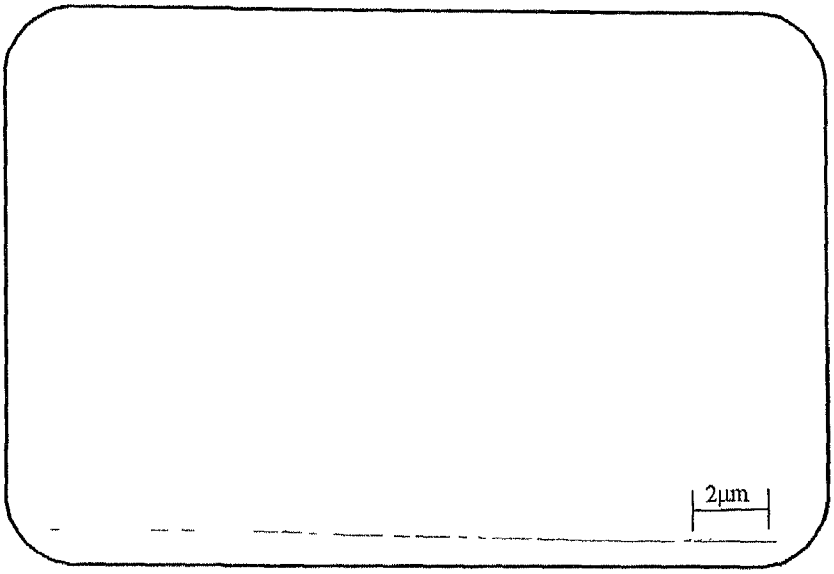


Figure 4 16 Microstructure of the sample rolled at 1100°C followed by TSC

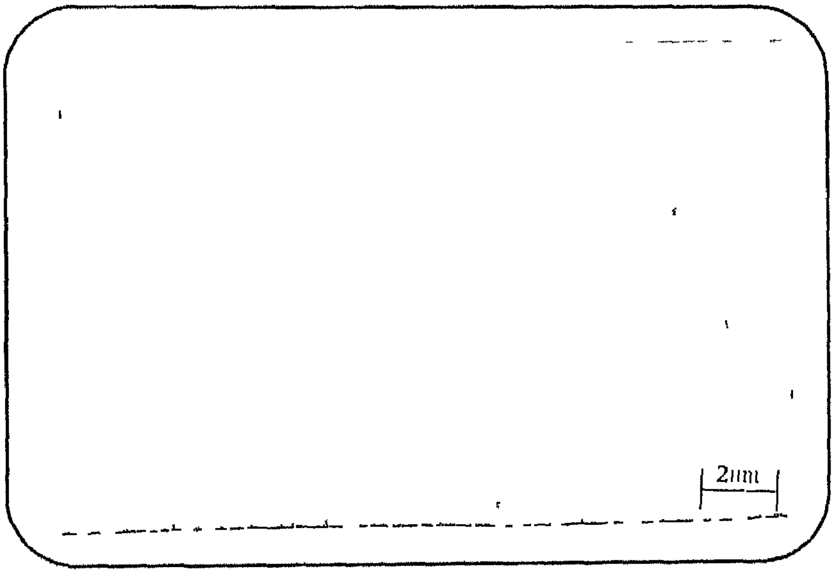


Figure 4 17 Microstructure of the sample rolled at 1100°C followed by TSC and annealed at 420°C for 1hr

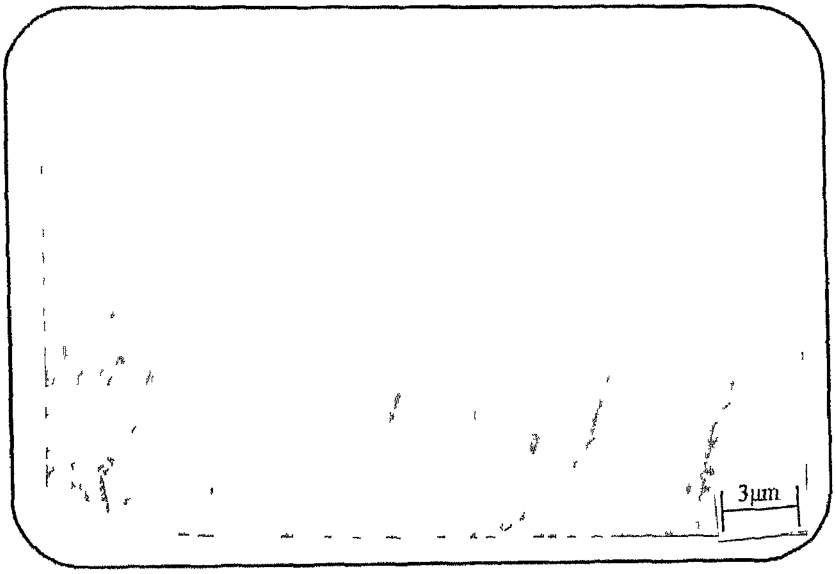


Figure 4 18 Microstructure of the sample rolled at 1200°C followed by TSC

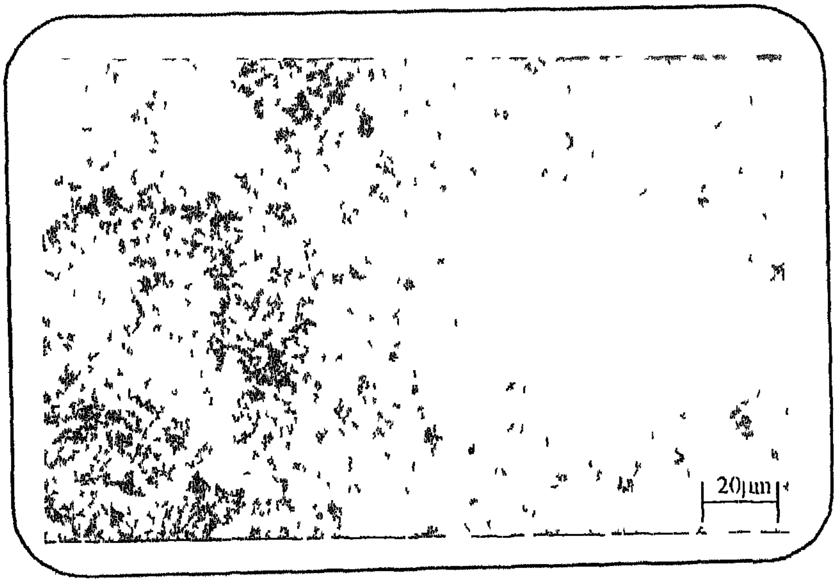


Figure 4 19 Microstructure of the sample rolled at 800°C followed by TSC

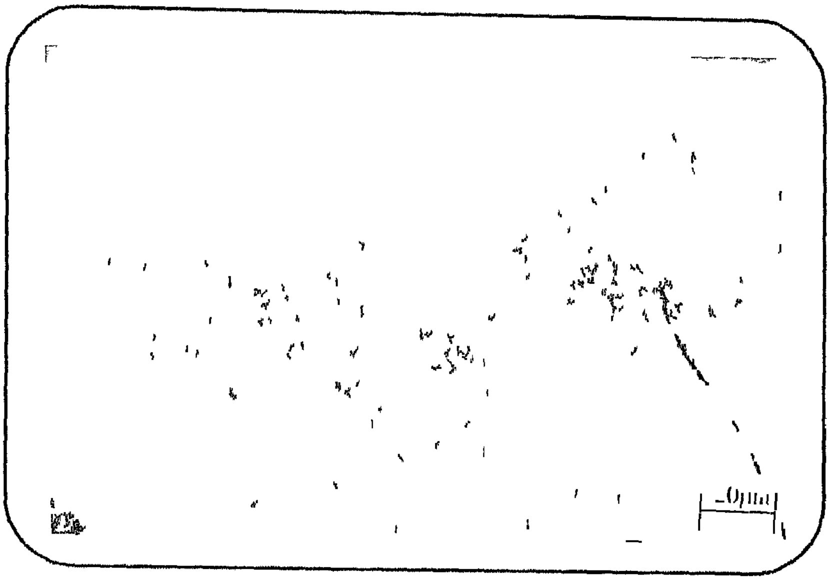


Figure 4 20 Microstructure of the sample rolled at 800°C followed by TSC and annealed at 420°C for 1hr

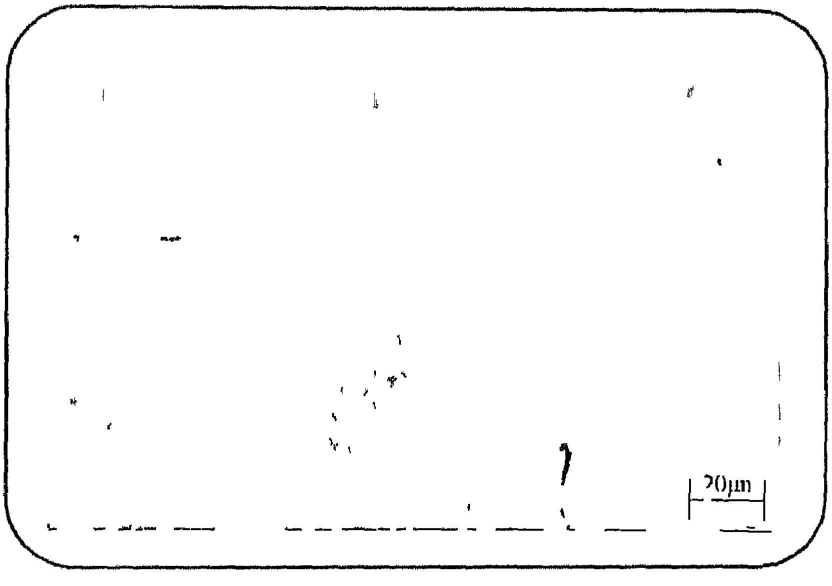


Figure 4 21 Microstructure of the sample rolled at 900°C followed by TSC and annealed at 420°C for 1hr

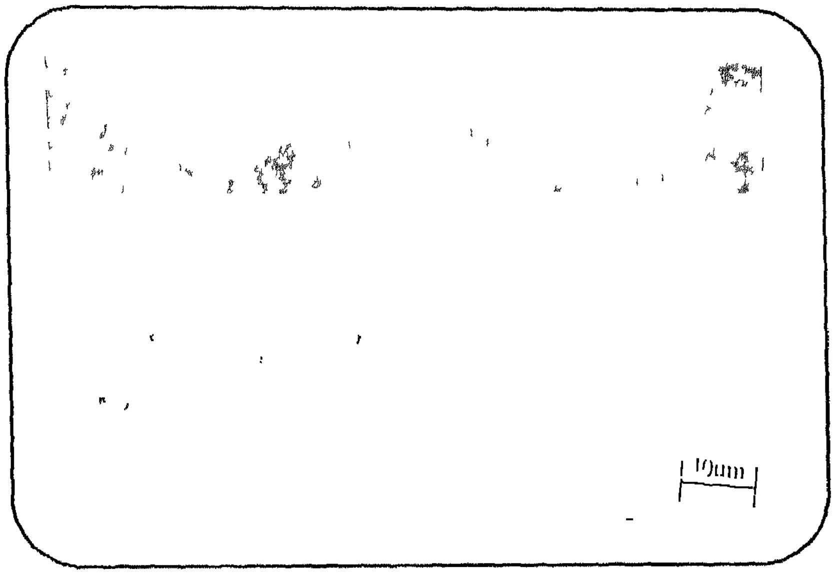


Figure 4 22 Microstructure of the sample rolled at 1000°C followed by TSC and annealed at 420°C for 1hr

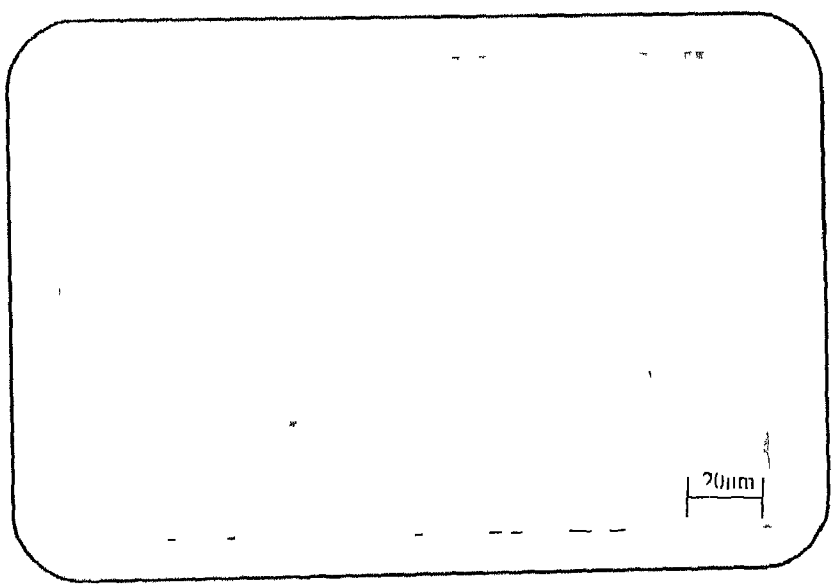


Figure 4 23 Microstructure of the sample rolled at 1100°C followed by TSC and annealed at 420°C for 1hr

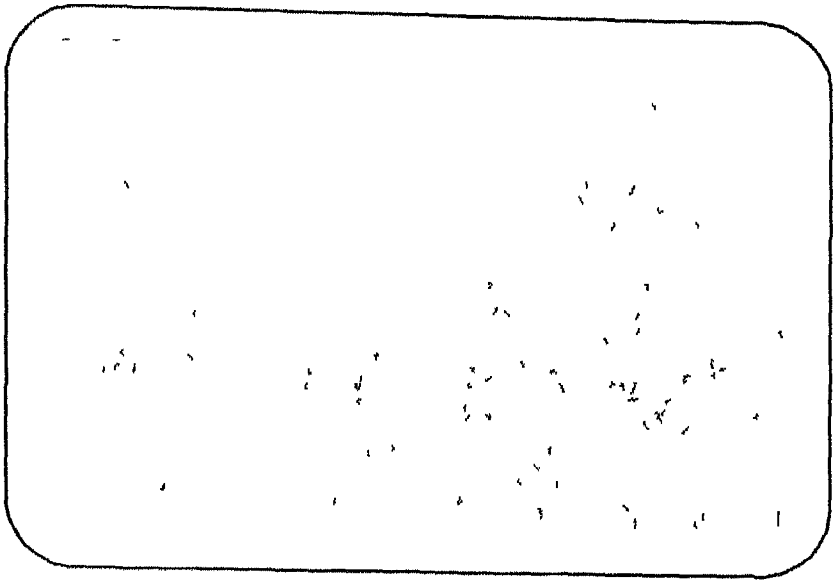


Figure 4 24 Microstructure of the sample rolled at 1200°C followed by TSC

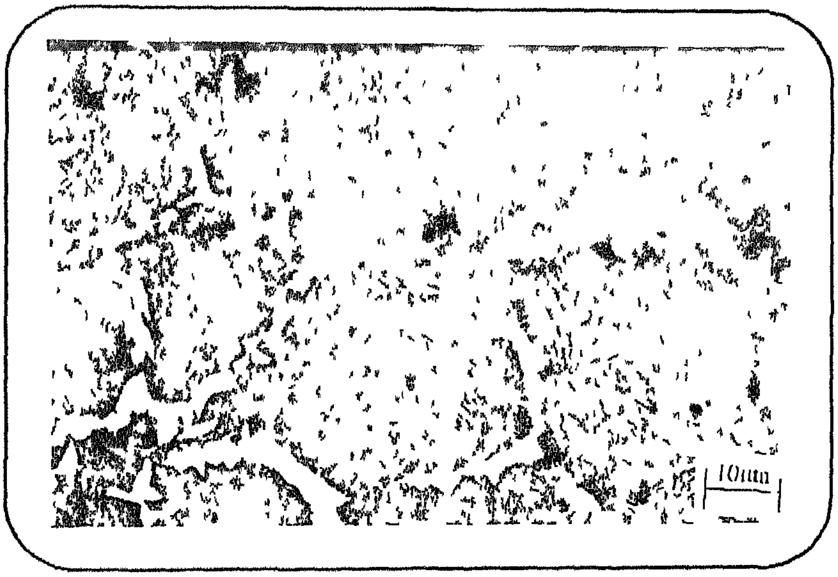


Figure 4 25 Microstructure of the sample rolled at 1200°C followed by TSC and annealed at 420°C for 1hr

4.2 Microstructural Features in Thermomechanically Treated 49MnVS3 Steel

4.2.1 Microstructures of Hot Rolled and Quenched Steel

Figure 4.2 is the SEM micrograph of the sample rolled at 760°C followed by water quenching. The Fe-Fe_3C phase diagram shows that the 49MNVS3 steel at this temperature consists of the two-phase mixture comprising of ferrite and austenite. The as-rolled and quenched microstructure of the steel, therefore comprises of a typical dual phase structure (Ferrite + Martensite).

Figures 4.3, 4.4, 4.6 and 4.7 are the micrographs of the samples rolled at single-phase field i.e. from 800°C, 900°C, 1000°C and 1100°C respectively. It may be seen that all the structures are typical martensitic structures. SEM micrograph (figure 4.4) shows the details of size of martensitic needles obtained in the sample quenched from the rolling temperature of 900°C. It was seen that the microstructure consisted of both long and short martensitic needles. As it is clear from figures 4.3, 4.6 and 4.7, the size of martensite needles increased with increasing the hot rolling temperature.

4.2.2 Microstructure of Hot Rolled Steel after Quenching and Tempering

In conventional quenching and tempering process the samples were rolled at different temperatures from 760° to 1200°C followed by water quenching and tempering/annealing at 560°C for 1 hr. Microstructures after tempering/annealing, in general, showed the mixture of a small amount of ferrite and tempered martensite. This small amount of ferrite is found to be forming in quenched and tempered steels containing 0.48 – 0.50 %C [Metals Handbook Vol 8]. This ferrite forms due to the tempering of retained austenite and decomposition of metastable martensite into the

more stable products of ferrite and cementite by the process of carbon diffusion during tempering The morphology of ferrite in the samples was, however, found to be influenced by the hot rolling temperature of the steel in the single phase austenitic field For example when the steel was rolled at 900°C, (Micrograph 4 5) small amount of ferrite was seen at the prior austenite grain boundaries with emerging spines of ferrite in the matrix of tempered martensite On the other hand, when the steel was rolled at higher temperatures, followed by quenching and tempering/annealing, ferrite was found to appear within the matrix only (Micrograph 4 8, hot rolling temperature = 1200°C)

4 2 3 Microstructures of Hot Rolled Steel after Two Step Cooling (TSC)

The two stage cooling process has recently been successfully adopted by Kothe and workers [Kasper et al, 1997] In such a process, the first slow cooling step (done in still air) produces certain fraction of soft proeutectoid ferrite Subsequently, during the accelerated cooling step (done by water quenching) gives rise to the bainite/martensite transformation in the remaining matrix

Figure 4 9 shows the SEM miscograph of the steel rolled at 760°C and cooled through TSC This microstructure was found to be a typical dual-phase microstructure and comprised of ferrite and martensite It must be noted that the microstructure of the steel hot rolled at 760°C and processed through TSC resembles that of hot rolled and quenched microstructure (shown in figure 4 2) The difference it has with the rolled and quenched microstructure is in terms of the volume fraction of ferrite which was found to be somewhat higher in the TSC steel The slightly higher volume fraction of ferrite arises due to its formation during the first step of the TSC when the steel cools in air 660°C

Microstructures of the steel rolled in the single-phase austenite field and subsequently cooled through TSC are shown in figures 4 10 (800°C), 4 11 and 4 12 (900°C), 4 13 (1000°C), 4 16 (1100°C) and 4 18 (1200°C) While the structure of the steel hot rolled at 800°C and cooled through TSC (figure 4 10) consisted of about 24 volume % of polygonal ferrite and predominantly upper bainite (acicular structure) in the matrix, that of the steel rolled at 1200°C and cooled through TSC (figure 4 18) consisted of about 19 volume % of ferrite and predominantly martensite in the matrix For the hot rolling temperatures in between these two extremes, i e for the rolling temperatures of 900°C, 1000°C and 1100°C, the volume fraction of ferrite fell continuously (figure 4 26) and the matrix structure changed from predominantly bainite to predominantly martensite Another microstructural feature that changed with the hot rolling temperature was the morphology of the ferrite While the ferrite morphology was found to be almost polygonal in the steel rolled at 800°C, it comprised of polygonal plus elongated shapes for steels rolled at higher temperature Figures 4 12 and 4 13 show the SEM micrographs of the sample rolled at 900°C and 1000°C respectively followed by TSC Both the structures consist of ferrite at the prior austenite grain boundaries with emerging spines of ferrite in the matrix of martensite So from the above discussion it was concluded that microstructure of the TSC samples consisted of softer ferrite at the prior austenite grain boundaries with emerging spines of ferrite in the matrix of harder martensite and/or bainite

4.2 4 Microstructures of Hot Rolled Steel after TSC and Annealing

The additional annealing step of heating at 420°C for one hour after the two stage cooling of the steel from the hot rolling temperature had an influence on the microstructure of 49MnVS3 steel Optical micrographs of TSC and annealed samples,

rolled at different temperatures, are shown in figures 4 20 - 4 25 While the matrix in all the conditions comprised of bainite and/or tempered martensite, the morphology of the ferrite phase was found to be differing with the rolling temperature Figures 4 20 and 4 21 show the optical micrographs of the steel rolled at 800°C and 900°C cooled through TSC and annealed It is seen that both these samples comprised of predominantly polygonal ferrite sitting at the prior austenite grain boundaries Similarly the microstructure of the steel rolled at 1200°C, cooled through TSC and annealed, is shown in figure 4 25 An almost continuous network of polygonal ferrite along prior austenite grain boundaries may be seen in this microstructural state In contrast, the structure of steel rolled at 1000°C and 1100°C consisted of most of the ferrite at the prior austenite grain boundaries and small amounts of ferrite within the matrix of tempered martensite/bainite It must be further noted that the ferrite within the matrix of tempered martensite/bainite was elongated and was in the spine-like morphology From the optical micrograph it is also clear that prior austenite grains were more equiaxed (figure 4 25) of the sample rolled at higher temperature (1200°C) than the sample rolled at lower temperature (Fig 4 21)

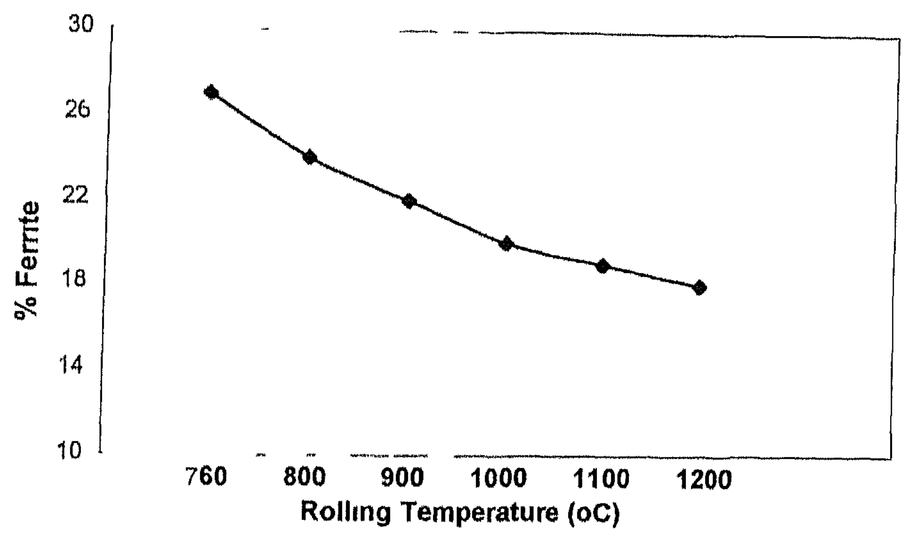


Figure 4 26· Variation of volume fraction of ferrite with rolling temperature

4.3 Mechanical Properties of Thermomechanically Treated 49MnVS3 Steel

4.3.1 Tensile Properties

The tensile tests were performed to understand the effect of microstructural changes in 49MnVS3 steel on its various tensile properties at room temperature, viz the yield strength, the ultimate tensile strength and the ductility The tensile properties of the steel were obtained for the following thermomechanically treated samples

- Hot rolled and water quenched samples,
- Hot rolled and water quenched samples followed by annealing at 560^0C,
- Hot rolled and cooled through TSC, i e air cooling up to 660^0C followed by water quenching,
- Hot rolled and cooled through TSC followed by an additional annealing at 420^0C

This section of the present chapter describes the results on the 0 2% yield strength, ultimate tensile strength and the % elongation obtained from flat tensile samples

Yield Strength:

0 2% yield strength of thermomechanically treated 49MnVS3 steel under different processing routes is shown in Table 4 1 and Figure 4 27 The yield strength for the steel rolled at 800^oC to 1200^oC and water quenched was not measured due to its martensitic structure and hence the ensuing brittleness It has therefore not reported in Table 4 1 However, as the steel processed through other routes, i e (a) hot rolling + quenching + tempering/annealing, (b) hot rolling + two stage cooling and (c) hot

Table 4 1 Comparison of Yield Strength of 49MnVS3 Steel Processed through Different Thermomechanical Routes

Hot Rolling Temperature (°C)	W.Q. (Mpa)	W Q +Annealing at 560°C(1h) (Mpa)	T.S C (Mpa)	T S C + Annealing at 420°C(1h) (Mpa)
760	575	550	507	549
800	*	905	904	1008
900	*	920	917	1033
1000	*	925	927	1034
1100	*	939	914	989
1200	*	914	923	990

* The yield strength has not been reported because these samples were essentially martensitic and displayed considerable brittleness (See Table 4 3) The yield point and the UTS almost coincided for these samples

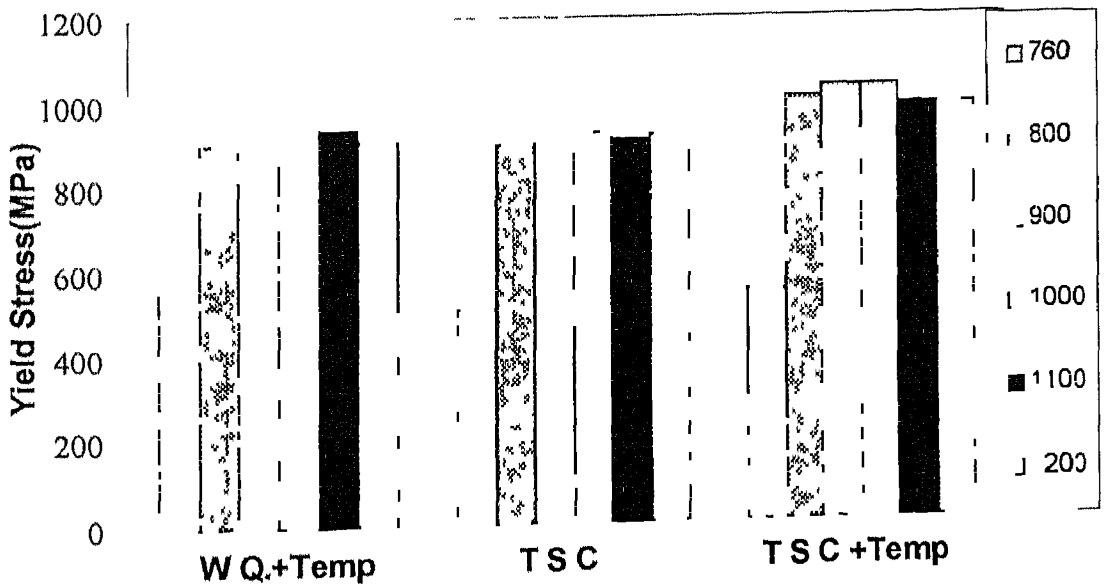


Figure 4 27. Variation of Yield strength with different Thermomechanical treatment of 49MnVS3 steel

rolling + two stage cooling + annealing, did not show brittleness, 0 2% yield strength was obtained for these conditions from the tensile tests

In general, the yield strength of the steel processed through a given thermomechanically treatment showed a maximum with the hot rolling temperature The maxima in each case occurred in the temperature range of 900°C - 1100°C and was definitely lower for the samples rolled at the temperature of 800°C Further, the highest yield strength in the steel was obtained in the TSC + annealed stage From the figure 4 27, it can be seen that yield strength is quite low of steel when rolled at 760°C It was also seen that yield strength of TSC sample is inferior that of quench and tempered variety but after additional annealing of TSC samples, the yield strength is about 10%higher than quenching and tempered variety

UTS

Ultimate tensile strength of thermomechanically treated 49MnVS3 steel under different processing routes is shown in Table 4 2 and Figure 4 28 From the figure 4 28, it was seen that there was large decrease in UTS after tempering of the water quenched sample It was also seen that UTS is about 15% to 18% higher in the steel under TSC condition than under quenched and tempered steel There was a little decrease in UTS after annealing of the steel cooled through TSC route due to the tempering of martensite/bainite It was also seen that the highest UTS in the steel was obtained when sample rolled at temperature between 900°C to 1000°C and cooled by TSC route

Table 4.2 Comparison of Ultimate tensile strength of 49MnVS3 Steel Processed through Different Thermomechanical Routes

Hot Rolling Temperature (°C)	W Q (Mpa)	W Q + Annealing at 560°C(1h) (Mpa)	T S.C (Mpa)	T S C + Annealing at 420°C(1h) (Mpa)
760	869	710	826	765
800	1893	998	1341	1272
900	1823	952	1457	1383
1000	1739	1015	1470	1371
1100	1861	1035	1243	1218
1200	1792	1143	1247	1212

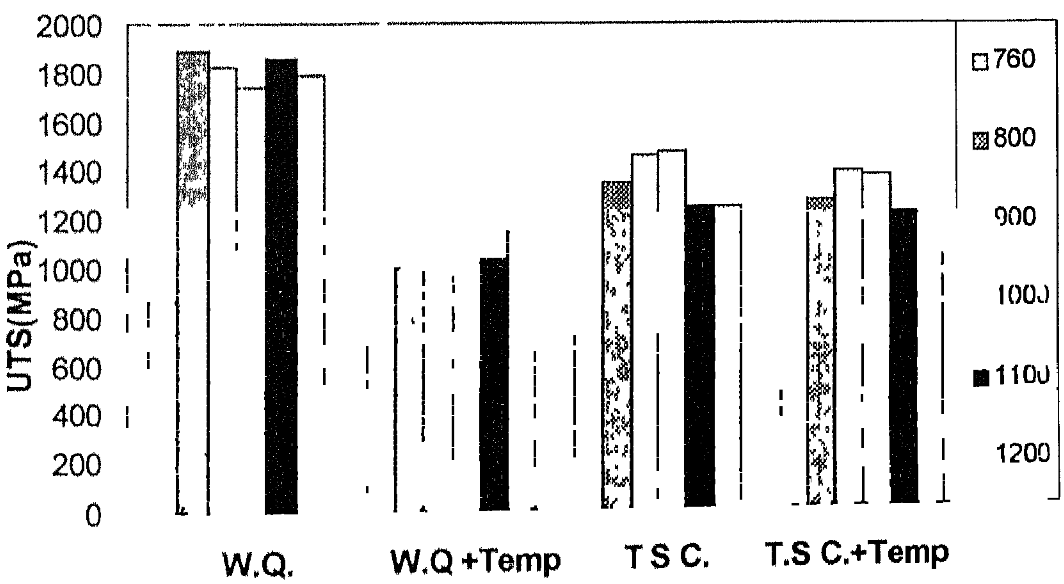


Figure 4.28: Variation of UTS with different Thermomechanical treatment of 49MnVS3 steel

Tabte 4 3 Comparison of % Elongation of 49MnVS3 Steel Processed through Different Thermomechanical Routes

Hot Rolling Temperature (°C)	W Q	W Q + Annealing at 560°C(1h)	T.S C	T S.C + Annealing at 420°C(1h)
	(%)	(%)	(%)	(%)
760	13 6	20 5	17 9	22 6
800	2 7	12 1	7 9	11 6
900	2 3	13 5	8 3	12 2
1000	1 9	13 3	8 7	12 7
1100	2 4	12 9	8 1	12 1
1200	2 1	11 9	7 7	11 9

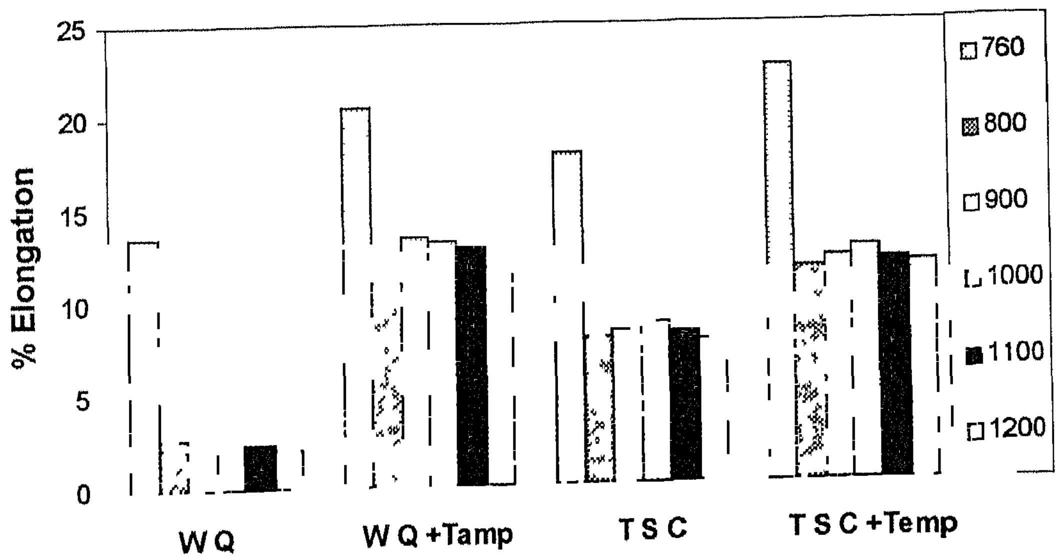


Figure 4 29. Variation of % Elongation with different Thermomechanical treatment of 49MnVS3 steel.

Elongation

Elongation of the steel under different processing routes is shown in Table 4 3 and Figure 4 29 From figure 4 29, it is clear ductility is quite high of the sample rolled at two-phase region (760°C) The ductility of the steel under TSC condition is inferior than that under quenches and tempered condition After annealing of the TSC sample, the percentage elongation increases upto 4-5% due to the tempering of martensite and/or bainite The ductility of the steel under TSC + Tempered condition is slightly lower than that under quenched and tempered condition when sample rolled in single phase field i e between 800°C to 1200°C

4 3 2 Toughness

The present investigation involved with the hot rolling of 49MnVS3 steel at different rolling temperatures, cooling it under different schemes and subsequently tempering/ annealing the samples The hot rolling was done on flat plate samples that had their starting thickness of ~ 7 mm These plates were reduced in thickness by giving ~ 50% thickness reduction so as to give the final thickness of ~ 3 5 mm Rolling of thicker plates was not possible due to the constraints of the given rolling set-up From the rolled material of ~ 3 5 mm thickness it was not possible to prepare Charpy/Izod samples for tough ness testing or the notched samples the testing of K_{Ic} value determination of the rolled steel samples Therefore, in order to have a qualitative understanding regarding the toughness of the thermomechanically treated 49MnVS3 steel, the area under the stress strain curve was measured to get the toughness of various thermomechanically treated steel samples

Table 4 4 and figure 4 30 show the variation of toughness of the differently treated materials From figure 4 30 it was seen that

Tabte 4 4 Comparison of Toughness of 49MnVS3 Steel Processed through Different Thermomechanical Routes

Hot Rolling Temperature (°C)	W Q	W Q + Annealing at 560°C(1h)	T S C	T S C + Annealing at 420°C(1h)
	(MJm^{-3})	(MJm^{-3})	(MJm^{-3})	(MJm^{-3})
760	101	137	119	138
800	*	115	89	132
900	*	126	99	144
1000	*	129	104	151
1100	*	127	88	134
1200	*	122	85	131

* Not measured because of very limited ductility of these samples

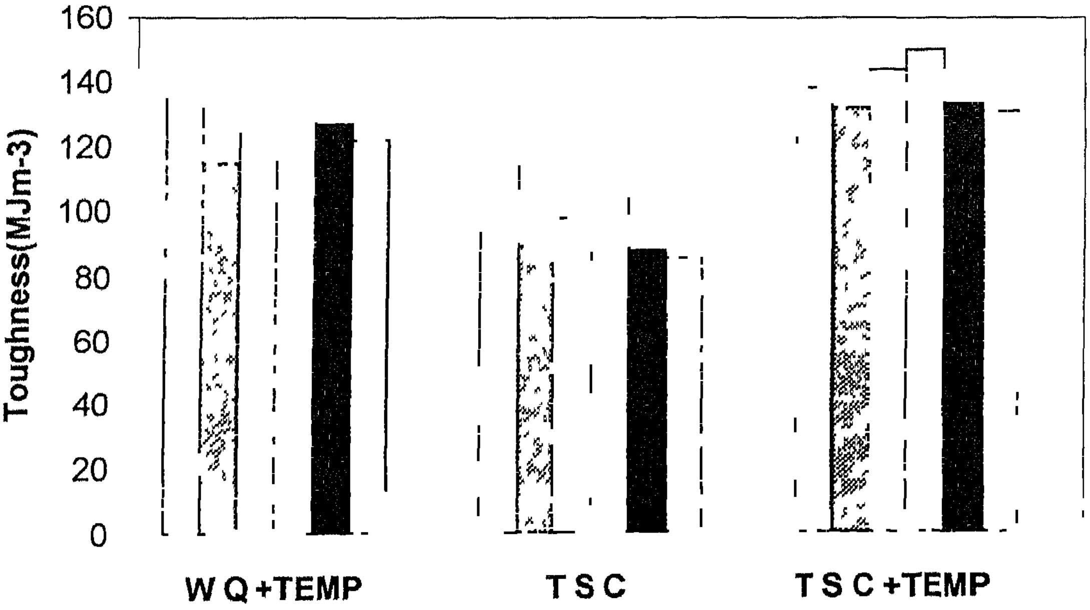


Figure 4 30 Variation of Toughness with different Thermomechanical treatment of 49MnVS3 steel

toughness of steel under TSC condition is inferior than that under quench + tempered condition On the other hand when due to the additional annealing of TSC sample, toughness increases up to 40% to 50% It was also seen that the maxima in each route occurred in the temperature range of 900°C - 1000°C when steel rolled at single phase field From figure it is clear that toughness of the steel rolled at 760°C (two phase field) is superior than the steel rolled at single phase field in the case of WQ + Tempered and TSC route On the other hand highest toughness value was obtained when the steel rolled between 900°C to 1000°C under TSC + annealed condition

4.3 3 Microhradness

It has already been discussed that the thermomechanical treatment of 49MnVS3 steel comprised of (I) hot rolling at different rolling temperatures followed by water quenching and subsequent tempering/annealing or (II) hot rolling at different rolling temperatures followed by a two stage cooling and subsequent annealing While the microstructures obtained in the first route comprised of ferrite and the tempered martensitic structure, those obtained in the second route comprised of ferrite plus bainite and/or tempered martensite Any quantitative analysis of volume fraction of bainite + martensite or tempered martensite in microstructures obtained through the second thermomechanical treatment route was found to be an extremely difficult task The microhardness measurement of the matrix was therefore undertaken to qualitatively assess the relative volume fractions of the bainite and martensite or tempered martensite phases in the structures of steel processed through TSC However, for the purpose of comparison, microhardness measurements were done on all the samples Due to the extremely fine size of the ferrite phase in all the processed samples, no microhardness measurements could be made on ferrite

Tabte 4 5 Comparison of Microhardness of 49MnVS3 Steel Processed through Different Thermomechanical Routes

Hot Rolling Temperature (°C)	W.Q.	W Q + Annealing at 560°C(1h)	T S.C.	T S.C + Annealing at 420°C(1h)
	(VHN)	(VHN)	(VHN)	(VHN)
760	351	286	296	262
800	891	423	841	412
900	874	423	857	441
1000	857	423	841	460
1100	841	441	810	501
1200	841	441	841	549

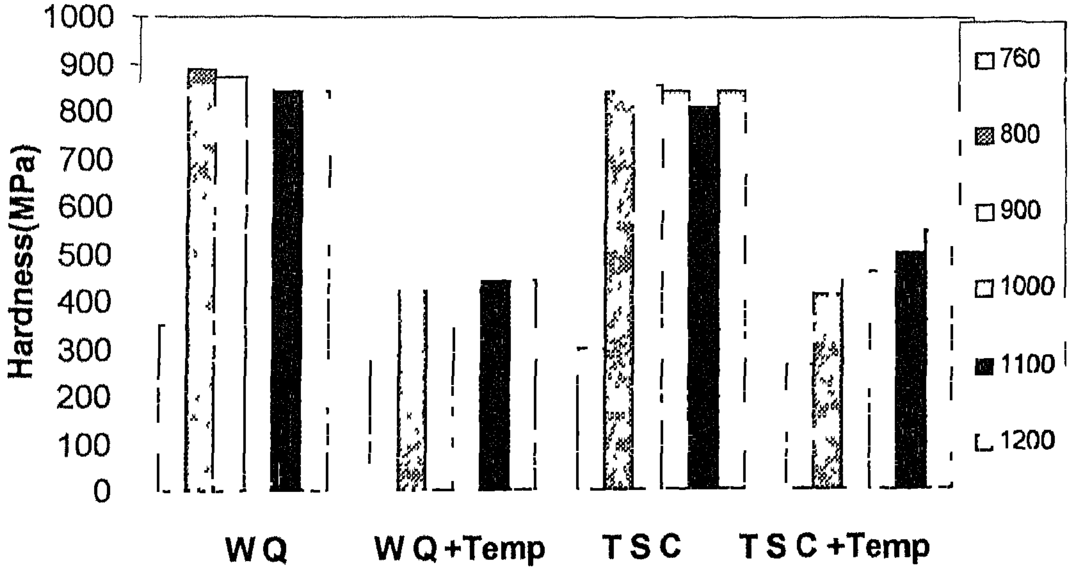


Figure 4 31· Variation of Microhardness with different Thermomechanical treatment of 49MnVS3 steel

Microhardness values of the martensitic/ tempered martensitic matrix (for the first thermomechanical treatment) and bainitic/martensitic/tempered martensitic matrix (for the second thermomechanical treatment) are shown in Table 4 5 and Figure 4 31 From the figure it was seen that microhardness is almost independent of temperature under W Q , W Q + Tempered and TSC conditions On the other hand hardness increases with rolling temperature under TSC + annealed conditions

4.4 Microstructure – Property Corelations in Thermomechanically Treated 49MnVS3 Steel

4 4 1 Microstructural Evolution in 49MnVS3 Steel during its Thermomechanical Treatment

The chemical composition of 49MnVS3 steel has been shown in Table 3 1 It can be seen that it is microalloyed medium-carbon steel Microstructural parameters of interest for manipulating its mechanical properties of microalloyed steels, such as the yield strength, the tensile ductility, the toughness and the rate of the fatigue crack growth, will be

- Prior austenite grains and their conditioning,
- Grain size of prior austenite grains,
- The volume fraction of carbide(s)/carbonitride(s) of the microalloying element(s), their size and size distribution,
- The amount of ferrite, its morphology, its size and the size distribution,
- Volume fractions of bainite and/or tempered martensite as matrix in the microstructure

Results shown in Section 4 2 of this chapter clearly show that most of these micro-structural parameters were influenced by the thermomechanical treatment

schedules followed in the present investigations Physical metallurgical considerations indicating that the thermomechanical treatment of 49MnVS3 steel will influence the above microstructural features are discussed below

The dissolution of carbide(s)/carbonitrides in the matrix of microalloyed steels is known to play an important role in the microstructural evolution of these steels during their thermomechanical treatment [Peter H Wright, 1990] First, if the carbide(s)/carbonitride(s) are not completely dissolved in the steel at the rolling temperature, the mechanical deformation occurs essentially of the austenite grains that are dispersed with carbide(s)/carbonitride(s) In such a case the carbon % of the austenite remains somewhat lower and the recrystallization temperature of the steel also remains low Second, when the carbide(s)/carbonitride(s) are completely dissolved in austenite, the deformation in the steel occurs mainly of the single phase, the austenite has a slightly higher carbon content and the recrystallization temperature of the steel is higher The dissolution behavior of vanadium carbide has been studied using thermodynamic considerations by several authors [Peter H Wright, 1990] Precipitation and dissolution characteristics of vanadium and niobium carbides in austenite have been shown in Figure 4 32

In the case of two stage cooling, rolling and quenching temperature is also play an important role in the microstructural evolution of these steels In TSC route, since all the samples were quench from 660°C, that's why volume fraction of ferrite and their shape size and size distribution depends on rolling temperature From the figure 4 26 it can be seen that volume fraction of ferrite decreases with increasing rolling temperature This can be explained with respect to ferrite start temperature Increasing the deformation temperature (T_d) leads to lowering the ferrite start temperature f_{s1} (figure 4 33) [Kothe et al , 1997], brings about a smaller time delay t_1 in the region of

ferrite formation i e higher is the deformation temperature smaller is the ferrite volume fraction

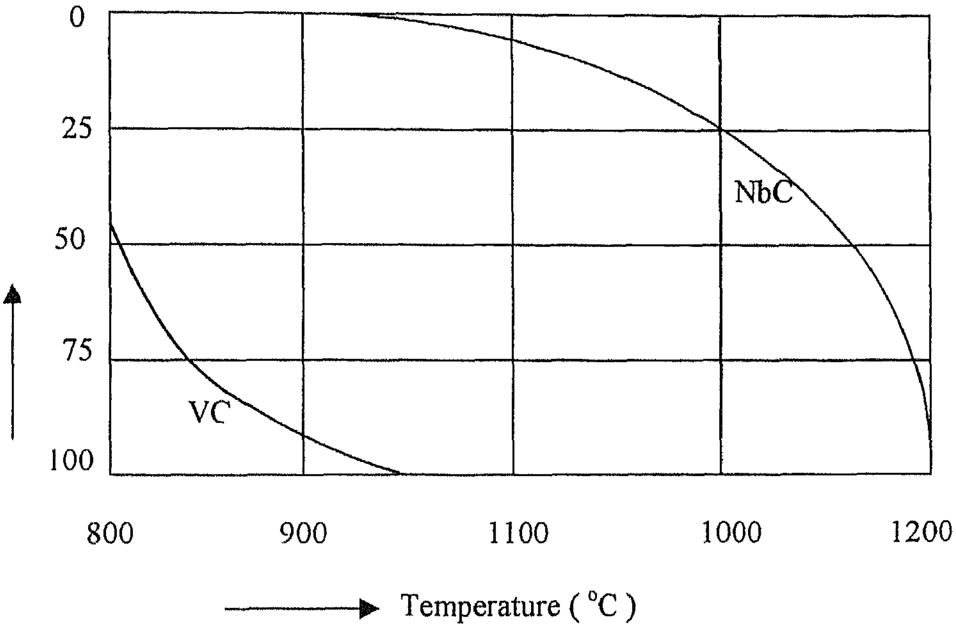


Figure 4 32 Precipitation and dissolution characteristics of vanadium and niobium carbides in austenite

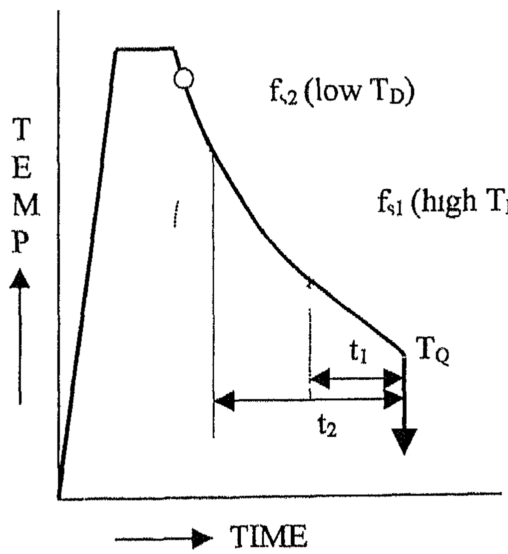


Figure 4 33 Changes in the ferrite formation time (between ferrite start temperature f_s and quenching temperature T_Q) due to the variation of deformation temperature

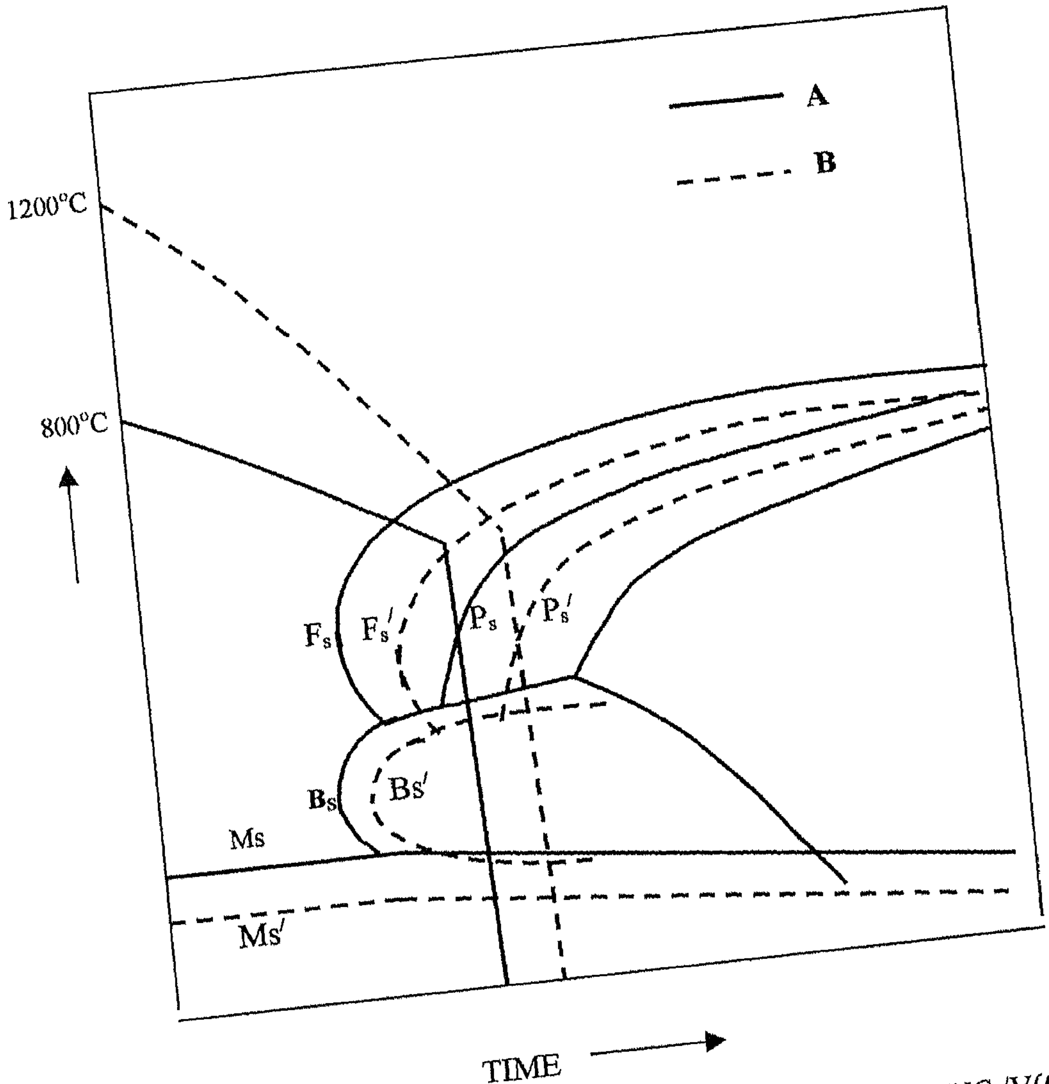


Figure 4 34 Schematic TTT diagram of 49MnVS3 steel (A) When VC /V(CN) partially dissolved in austenite (B) when VC/V(CN) completely dissolved in austenite

The steel hot rolled at 800°C and cooled through TSC (figure 4 10) consisted of about 24 volume % of polygonal ferrite and predominantly upper bainite (acicular structure) in the matrix, but steel rolled at 1200°C and cooled through TSC (figure 4 18) consisted of about 19 5 volume % of ferrite and predominantly martensite in the matrix For the hot rolling temperatures in between these two extremes, i e form the rolling temperatures of 900°C, 1000°C and 1100°C, the volume fraction of ferrite fell continuously (figure 4 26) and the matrix structure changed from predominantly bainite to predominantly martensite This can be explained from figure 4 34

Figure 4 34 represents the TTT diagram (schematic) of 49MnVS3 steel Below 900°C VC/V(CN) is not completely dissolved in austenite(figure 4 32) So at low temperature the amount of dissolved carbon is lower than the total carbon of the same steel Due to the lower carbon at low temperature the TTT curve shifts to the left and martensite start temperature(Ms) increases On the other hand at high temperature (above 950°C) VC\V(CN) completely goes in to solution So at high temperature the amount of carbon increases which shift the TTT curve to the right and Ms temperature also decreases In the figure 4 34, (A) continuous line represents the TTT curve correspond to steel rolled at low temperature and (B) hatch line represents the TTT curve correspond to steel rolled at high temperature Higher Ms temperature and left sift of TTT curve reduce the hardenability of steel and promotes the formation of bainite instead of martensite during two step cooling On the other hand right sift of TTT curve increase the hardenability From the figure it is seen that due to the lowering of Ms temperature there is a gap between Bs curve and Ms line Above two factors promotes the formation of martensite when steel rolled at high temperature and cooled through TSC route

4.4.2 Effect of Microstructure on Mechanical Properties of 49MnVS3 Steel

From the figure 4.26, it can be seen that yield strength is quite low of steel when rolled at 760^{0}C. The Fe-Fe_3C phase diagram shows that the 49MNVS3 steel at this temperature consists of the two-phase mixture comprising of ferrite and austenite. This is due to presence of large volume fraction of softer phase (ferrite) in the microstructure. From the figure 4.27 it is also clear that there is almost no effect of rolling temperature (about 800^{0}C) on yield strength. This can explain with respect to the similar microstructure i.e., the volume fraction of hard and soft phases are almost same in all the samples. It was also seen that yield strength of TSC sample is inferior that of quench and tempered variety but after additional annealing of TSC samples, the yield strength is about 10%higher than quenching and tempered variety. The increase in yield strength with additional annealing can be explained with respect to the precipitation of Vanadium carbide and Vanadium carbonitride during annealing. Vanadium carbides and carbonitrides preferentially precipitating in the ferrite from supersaturated ferrite during annealing. Which cause secondary hardening. The lower yield strength of TSC sample is due to the production of mobile dislocations in the ferrite, close to the ferrite-martensite interphase, during the austenite to martensite transformation[kaspar et al., 1997] due to the volume expansion of martensite. This makes the onset of plastic deformation easier.

From the figure 4.28, it was seen that there was large decrease in UTS after tempering of the water quenched sample due to the tempering of martensite and/or bainite. It was also seen that UTS is about 15% to 18% higher in TSC sample that of quenched and tempered sample. The higher strength of the TSC sample due to presence of multiphase microstructure (combination of hard untempered martensite/bainite with soft ferrite). There was a little decrease in UTS after annealing

of the TSC sample due to the tempering of martensite/bainite It was also seen from figures 4 27 and 4 28, sampled rolled at temperature between 900°C to 1000°C have the better yield stress and UTS This can be explained with respect to variation of precipitation strengthening effect of the sample rolled at different temperature Increasing the deformation temperature (T_d) leads to lowering the ferrite start temperature f_{s1} (figure 4 34) [kothe et al , 1997], brings about a smaller time delay t_1 in the region of ferrite formation and so a lower amount of precipitation of vanadium carbide (VC) or vanadium carbonitride V(CN) in ferrite Figure 4 32 [Wright, 1990] shows dissolution characteristics of vanadium carbide (VC) in austenite From this figure it was seen that the dissolution of vanadium carbide below 900°C is not completed So Yield Stress and UTS of the sampled rolled below 900°C is inferior due to less precipitation effect of VC or V(CN) in the austenite

From figure 4 29, it is clear ductility is quite high of the sample rolled at two-phase region (760°C) The higher ductility of the sample is due to the larger volume fraction of softer phase (ferrite) The ductility of the TSC samples is inferior than that of quenches and tempered sample This can be explained with respect to untempered martensite or bainite microstructure of the TSC sample After annealing of the TSC sample, the percentage elongation increases upto 4-5% due to the tempering of martensite or bainite The ductility of the TSC+ Tempered sample is slightly lower than the quenched and tempered sample due to the lower tempering temperature (420^{0}C) of the TSC sample The TSC sample was annealed at 420^{0}C because the strength, ductility combination is superiors at temperature 420^{0}C[kothe et al , 1997]

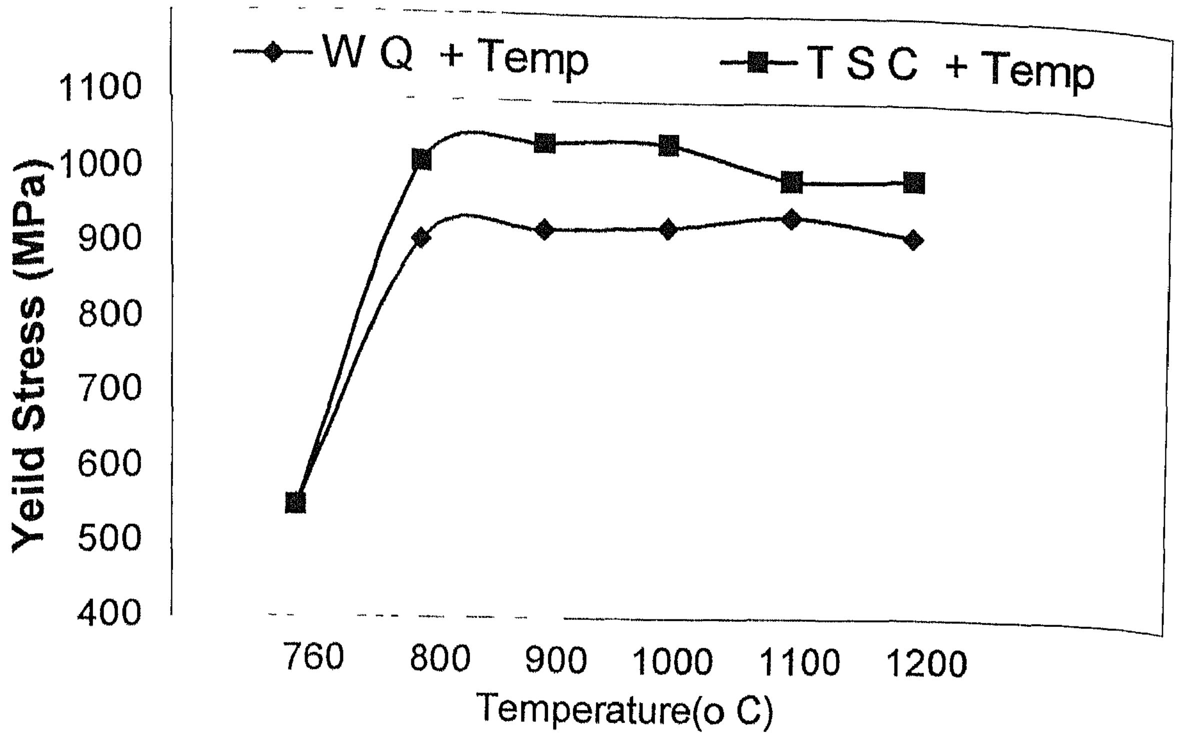


Figure 4.35 Comparison of Yield Strength of 49MnVS3 Steel Processed through Different Thermomechanical Routes

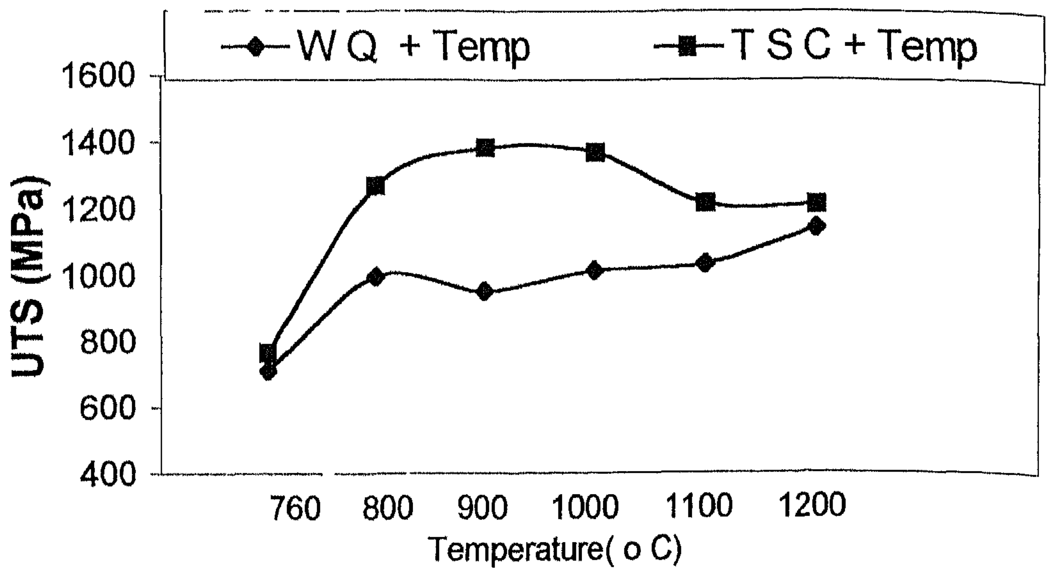


Figure 4.36 Comparison of Ultimate tensile strength of 49MnVS3 Steel Processed through Different Thermomechanical Routes

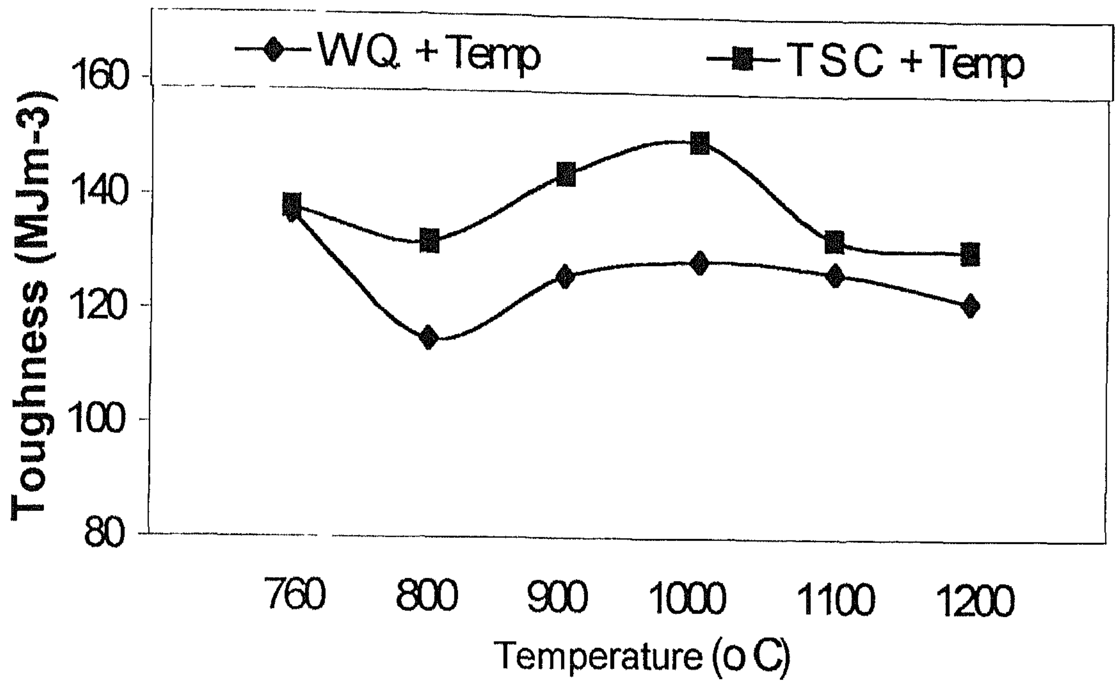


Figure 4 37 Comparison of Toughness of 49MnVS3 Steel Processed through Different Thermomechanical Routes

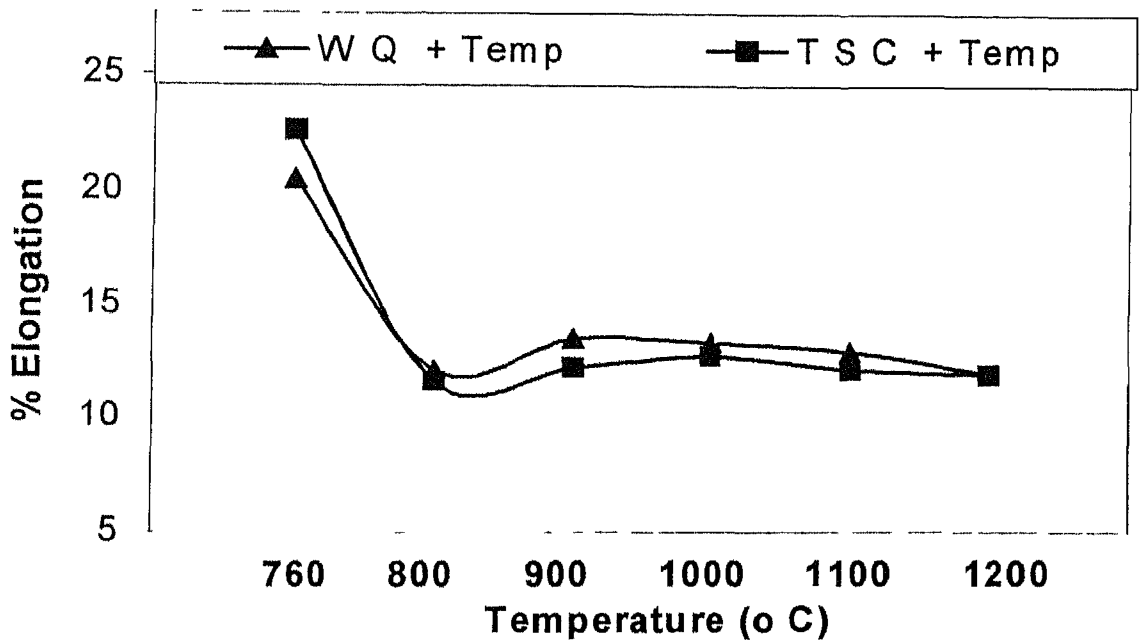


Figure 4 38 Comparison of % Elongation of 49MnVS3 Steel Processed through Different Thermomechanical Routes

From the figure 4 35-4 38, it is clear that ductility of the steel under TSC + annealed condition is almost same as that under the QT condition On the other hand, both yield strength as well as UTS are much higher of steel under TSC + annealed condition than that under the QT condition As a result the toughness of the steel under the TSC + annealed condition is 15%-16% higher than under the QT condition when rolled at the temperature between 900°C to 1000°C

CHAPTER 5

Conclusions

1 Thermomechanical treatments of 49MnVS3 steel play an important role in altering its microstructural state and hence its mechanical properties

2 The treatment involving hot rolling at 760°C followed by quenching provides a dual-phase structure comprising of polygonal ferrite and martensite After a tempering/ annealing treatment carried out at 560°C for 1 hour, the structure comprises of ferrite and tempered martensite However, under these conditions the steel shows a lower strength (550 MPa) than the highest achievable value in the same steel by the two stage cooling (TSC) route (1034 MPa) However, the ductility of the steel under this state is considerably superior (~ 20 5 % tensile elongation) when compared to TSC route (~ 12 7 % tensile elongation)

3 The route involving hot rolling of the steel in the single phase austenite field in the temperature range from 800°C – 1200°C followed by the quenching and tempering/ annealing (QT) does not provide much variation in microstructures and hence the consequent properties Under this condition, however, the steel shows a superior yield strength (~ 939 MPa) than that obtained in the steel rolled in the two-phase field (550 MPa) but a lower ductility

4 The route involving hot rolling of the steel in the single phase austenite field in the temperature range from 800°C – 1200°C followed by two stage cooling (TSC) and tempering/annealing provides considerable variation in its microstructure and mechanical properties of the steel

5 Physical metallurgical considerations involving the two stage cooling after the hot rolling of steel at different temperatures suggest that microstructural variation in the steel may arise due to (a) grain size of prior austenite grains, (b) the volume fraction of carbide(s)/carbonitride(s) of the microalloying element(s), their size and size distribution, (c) the amount of ferrite, its morphology, its size and the size distribution, (d) volume fractions of bainite and/or tempered martensite as matrix in the microstructure All such variations in microstructures were found to be associated with the steel differently hot rolled and cooled through TSC and subsequently annealed at 420°C

6 The present investigation shows that the steel hot rolled at 800°C – cooled through TSC – annealed at 420°C consists of a small volume fraction of polygonal ferrite (~24 vol%) and bainite as the matrix In contrast, steel hot rolled at 1200°C – cooled through TSC – annealed at 420°C consists of a somewhat smaller volume fraction of polygonal ferrite (~ 19 vol%) and tempered martensite as the matrix The steel rolled in between these two temperatures, i e at 900°C, 1000°C and 1100°C, shows polygonal as well as elongated ferrite and a mixture of bainite and tempered martensite as the matrix The volume fraction of the bainite in the matrix, however, decreases as the rolling temperature of the steel increases

7 Ductility of the steel under TSC + annealed condition is almost same as that under the QT condition On the other hand, both yield strength as well as UTS are much higher of steel under TSC + annealed condition than that under the QT condition As a result the toughness of the steel under the TSC + annealed condition is 15%-16% higher than under the QT condition when rolled at the temperature between 900°C to 1000°C

8 Among the thermomechanical routes investigated in the present study, it has been found that the treatment involving a combination of hot rolling followed by two stage cooling and annealing is the best thermomechanical route for the processing of 49MnVS3 micro-alloyed steel from the point of view of optimizing its mechanical properties It was also observed the optimum mechanical properties are obtained of the steel under TSC + annealed condition and when rolled at temperature between 900°C to 1000°C

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